ALLOYS OF IRON RESEARCH, MONOGRAPH SERIES

FRANK T. SISCO, EDITOR

THE ALLOYS OF IRON AND CARBON

VOL. II—PROPERTIES

In accordance with War Production Board regulations, this book is printed in a war-time economy format with reduced margins. However, the type page is standard for a book of this kind, and ease of reading is maintained.

At the end of the war when paper and binding materials are again plentiful, this book and other similar volumes will be reprinted in larger size with normal margins.

ALLOYS OF IRON RESEARCH

Monograph Series

THE ALLOYS OF IRON AND MOLYBDENUM *by* J. L. Gregg

THE ALLOYS OF IRON AND SILICON *by* Earl S. Greiner, J. S. Marsh, and Bradley Stoughton

THE ALLOYS OF IRON AND TUNGSTEN *by* J. L. Gregg

THE ALLOYS OF IRON AND COPPER *by* J. L. Gregg and B. N. Daniloff

PRINCIPLES OF PHASE DIAGRAMS *by* J. S. Marsh

THE METAL—IRON *by* H. E. Cleaves and J. G. Thompson

THE ALLOYS OF IRON AND CARBON, VOL. I—CONSTITUTION *by* Samuel Epstein

THE ALLOYS OF IRON AND CARBON, VOL. II—PROPERTIES *by* Frank T. Sisco

THE ALLOYS OF IRON AND CHROMIUM, VOL. I—LOW-CHROMIUM ALLOYS *by* A. B. Kinzel and Walter Crafts

THE ALLOYS OF IRON AND CHROMIUM, VOL. II—HIGH-CHROMIUM ALLOYS *by* A. B. Kinzel and Russell Franks

THE ALLOYS OF IRON AND NICKEL, VOL. I—SPECIAL-PURPOSE ALLOYS *by* J. S. Marsh

In Preparation

THE ALLOYS OF IRON AND NICKEL, VOL. II—STEELS AND CAST IRONS

THE ALLOYS OF IRON AND VANADIUM

THE ALLOYS OF IRON AND MANGANESE

THE ALLOYS OF IRON AND CARBON

VOL. II—PROPERTIES

BY

FRANK T. SISCO

Metallurgist and Editor, Alloys of Iron Research

FIRST EDITION
THIRD IMPRESSION

Published for

THE ENGINEERING FOUNDATION

by the

McGRAW-HILL BOOK COMPANY, INC.

NEW YORK AND LONDON

1937

PRINTED IN THE UNITED STATES OF AMERICA

THE MAPLE PRESS COMPANY, YORK, PA.

ALLOYS OF IRON RESEARCH

MONOGRAPH SERIES

These monographs are a concise but comprehensive critical summary of research on iron and its alloys as reported in the technical literature of the world. They contain a discussion of all available data on binary and higher ferrous alloy systems, and on the effect of the alloying elements on carbon steel and on simple and complex alloy steels and special alloy cast irons. They provide a reliable foundation for further research and supply to the practical metallurgist, steel worker, foundryman, and engineer the essential information now scattered through more than two thousand journals and textbooks in many languages.

The authors are responsible for selection and evaluation of the data, for arrangement of subject matter, and for style of presentation. Each book, however, has been reviewed in manuscript by men especially qualified to criticize all statements. Indebtedness for this cooperation is recognized in the Acknowledgments. Finally, each manuscript has been reviewed and approved for publication by the Iron Alloys Committee.

The Committee expresses its appreciation to *The Engineering Foundation*, the *iron and steel, automotive and other industries of the United States and the American Iron and Steel Institute, Battelle Memorial Institute, The American Society for Metals, The American Foundrymen's Association, and National Bureau of Standards* for financial support, which made the laborious review of the world's literature possible; and to the libraries, engineering societies, and the technical press in the United States, Canada, England, Germany, France, Italy, Sweden, Japan, and Czechoslovakia, for cooperation in making available inaccessible reports and in permitting the use of published data.

669.9

PREFACE

This volume is the final portion of a correlation of data and a critical summary of the world's knowledge on the constitution and properties of iron-carbon alloys. The preparation of this book on the properties of carbon steels and cast irons has presented many problems not encountered in most of the other monographs of this series. One of the important problems was to decide what should be included and what should be omitted, in other words, whether this volume could be made to cover the field with reasonable completeness without assuming the bulk of an encyclopedia.

The primary object of Alloys of Iron Research is to prepare monographs which are comprehensive critical summaries of the world's research on iron and its alloys and which should, necessarily, contain a discussion of all important available data. This postulates that the literature be thoroughly reviewed and that nothing of importance be omitted from the bibliography and from mention in the text. In the case of the present book this was impossible. The extent of the literature on the properties of plain carbon steels and cast irons is almost boundless. Even after rejecting from the thousands of published articles the direct duplications, condensations, and abstracts and the papers which restate previously reported facts in slightly different form, more than two thousand papers remained. Although these were reviewed in preparing the manuscript, it was impossible to include even all the important ones and still keep the book to a reasonable size.

The procedure finally adopted for this second portion of the iron-carbon monograph was, therefore, the same as that used by Samuel Epstein in preparing Volume I, namely, to sift the literature carefully and to select those data which would establish most readily the desired base line of properties—a base line with which the properties of alloy steels may be compared—and which would indicate most clearly the engineering and other properties expected from commercial carbon steels and cast irons.

There are so many possible variations in the composition, structure, and treatment of commercial carbon steels and cast irons that it was not unexpected to find, when the literature was reviewed, that there were few if any data upon the effect of carbon as the only variable influencing the properties. It was, however, somewhat surprising to learn that the amount of data on the various subjects to be covered varied widely. Owing to the small amount of information available for some of these subjects—the properties of hot-rolled steel is a notable example—it was possible to review and correlate the published data with some degree of completeness. For other subjects there was such a mass of information available that the data finally selected are only a small fraction of the total amount published. In this case the selection was made, from papers available at one or more large libraries in this country, primarily to establish most clearly the desired line of reference, and not exclusively on the basis of merit. Consequently there are a number of excellent pieces of work which are not mentioned in this monograph. That they are not included is no reflection upon their quality; it means only that there are other reports, more readily available, which contain similar information or which establish the desired line of reference more readily. For still other subjects, particularly those in which the method of test has as much influence upon the properties as the material, the selection of data was made primarily to indicate the trend due to carbon and to give typical values for typical materials tested by the most readily reproducible methods.

As this monograph is fundamentally a base line for the monographs on the other alloys of iron, and as it is more likely to be used as a reference book than some of the others, all temperatures are given in degrees centigrade and, with the exception of a few in Chapters XV and XVI on physical constants, have been converted to the nearest 5 degrees Fahrenheit.

FRANK T. SISCO.

NEW YORK, N. Y.,
January, 1937.

ACKNOWLEDGMENTS

The preparation of a monograph on the alloys of iron and carbon was started six years ago at Battelle Memorial Institute as a part of the Institute's contribution to Alloys of Iron Research. The review of the literature, especially laborious and time consuming for plain carbon steels and cast irons, and the preparation of a first draft of the manuscript were undertaken by the Institute staff and were completed in 1934. Owing to the vast amount of data assembled, the Iron Alloys Committee considered it inadvisable to publish the monograph as a single volume. Hence the preliminary manuscript was divided into two parts: one on constitution and heat treatment and one on properties. The final manuscript for Volume I was written by Samuel Epstein, of Battelle Memorial Institute, and was published in June, 1936.

Using the large store of material on properties accumulated by the Institute staff and augmented by the data that appeared in the literature after the preliminary manuscript was completed, the author of the present volume correlated the data and wrote the final manuscript with the exception noted below. The original review of the literature and the original correlation of the data (by far the hardest job in the preparation of this monograph) as made by Battelle Memorial Institute were used by the author as he saw fit, and he is therefore wholly responsible for the selection and evaluation of the data, for the method of presentation, and for the accuracy of the conclusions. This applies to the entire book except Chapters XV and XVI, the final draft of which was prepared by J. S. Marsh, Physical Metallurgist of the Alloys of Iron Research Staff. A large part of the first draft of Chapters XI and XIV, revised in accordance with the suggestions of the reviewers, was used unchanged.

The author desires to acknowledge his indebtedness to the small army of metallurgists and others without whose help the book could hardly have been completed in a reasonable time. First and most important, he wishes to thank the members of the

staff of Battelle Memorial Institute, who reviewed the literature, and the 45 metallurgists named in the acknowledgments to Epstein's book, who reviewed the first draft prepared at Battelle and whose comments were so valuable in preparing the final manuscript.

Next he wishes to express his appreciation to those named below, who reviewed so conscientiously the second and following drafts and the final manuscript or who supplied unpublished data for inclusion in the monograph:

J. W. Bolton	The Lunkenheimer Company
G. A. Ellinger	National Bureau of Standards
A. B. Kinzel	Union Carbide and Carbon Research Laboratories
E. E. Legge	American Steel and Wire Company
D. J. McAdam, Jr.	National Bureau of Standards
P. E. McKinney	Bethlehem Steel Company
J. H. Nead	Inland Steel Company
C. M. Saeger, Jr.	National Bureau of Standards
H. A. Schwartz	National Malleable and Steel Castings Company
A. P. Spooner	Bethlehem Steel Company
R. L. Templin	Aluminum Company of America
J. S. Vanick	International Nickel Company
F. M. Walters, Jr.	Youngstown Sheet and Tube Company

The members of the Joint Research Committee of the American Society for Testing Materials and the American Society of Mechanical Engineers, on the Effect of Temperature on the Properties of Metals, reviewed Chapters XII and XIII, and the members of the Fatigue Research Committee of the American Society for Testing Materials reviewed Chapter XI. These chapters were revised in line with their authoritative comments.

All the members of the Iron Alloys Committee reviewed the several drafts of the monograph. Special indebtedness is expressed for the assistance rendered by J. T. MacKenzie in reviewing the chapters on cast iron and by John Johnston and the following members of his staff: Messrs. R. H. Aborn, E. S. Davenport, E. H. Engel, and J. R. Vilella.

FRANK T. SISCO.

NEW YORK, N. Y.,
January, 1937.

CONTENTS

THE ALLOYS OF IRON AND CARBON

VOLUME II

PROPERTIES

CHAPTER I

INTRODUCTION

Classes of Commercial Iron-carbon Alloys—Variables Affecting the Properties of Commercial Iron-carbon Alloys—Tensile Tests—Statistical Methods Applied to Mechanical Properties

Steel is civilization's most important material and owes its importance almost wholly to carbon. Its form, distribution, and behavior, when alloyed with iron, determine whether the iron will be malleable or not, strong or weak, tough or brittle, hard or soft. Since the beginning of the steel age, with increasing use of steam and the resulting economical methods for melting and refining, progress in making and adapting steel to man's needs has depended almost solely on his advancing knowledge of the rôle played by carbon. Even the development of alloy steels has not dimmed the importance of carbon; most alloying elements are added for the specific purpose of modifying the structure and improving the properties of a carbon steel. It is, of course, true that there are a few iron alloys, known commercially as alloy steels, in which more than a very small amount of carbon is undesirable, but in general all alloy steels may be considered as carbon steels containing a specific amount of one or more additional alloying elements. From this it follows that the best method of studying the structure and properties of any alloy steel

is to compare it with an unalloyed carbon steel of otherwise similar composition and of similar history.

The dual object of a monograph on the iron-carbon alloys was stated by Epstein in Volume I. A portion of this will bear repeating. "These commercial alloys are of interest to many—to metallurgist, steel maker, engineer, and a number of others—for two reasons. In the first place, the constitution and properties of carbon steels and cast irons serve as a base line with which the increasingly useful alloy steels may be compared. . . . In the second place, a summary of data on constitution and properties helps to establish the value of plain carbon steels and cast irons as structural materials and should indicate what engineering properties can be and are being secured without the use of alloying elements."

Before taking up the correlation of data on the properties of commercial iron-carbon alloys it is advisable, in this first chapter, to discuss briefly the various classes of iron-carbon alloys so that there will be no misunderstanding as to what is meant by carbon steel, cast iron, or malleable iron; to outline, also briefly, some of the variables which may have an effect on the properties; to define what is meant by tensile strength, yield strength, and other mechanical properties as used in this and other monographs of the series, properties which are not defined in more detail later in the book; and to advocate, tentatively and somewhat hesitantly—in view of the many other important things metallurgists must do before our knowledge of carbon steel and cast iron is complete—the wider use of statistical methods in studying the mechanical properties of commercial steels and cast irons.

A. CLASSES OF COMMERCIAL IRON-CARBON ALLOYS

In the 80 years which have elapsed since the commercial development of the Bessemer and open-hearth processes, metallurgists, chemists, engineers, and teachers have been attempting to define steel and to distinguish it by definition from ingot iron, wrought iron, cast iron, and other products of the iron and steel industry. Some 50 years ago, these efforts were complicated by the development of alloy steels which made it necessary to distinguish between ordinary steels in which carbon is theoretically the sole, and actually the main, controlling composition variable, and alloy steels in which the effect of the carbon is intensified and occasion-

ally partly nullified—but always complicated—by the effect of one or more alloying elements.

More than 60 years ago an international committee was appointed to consider the question; in 1876 a preliminary report was made to the American Institute of Mining Engineers. The work was continued until 1906 when the Committee on Uniform Nomenclature of Iron and Steel of the International Association for Testing Materials, with H. M. Howe as chairman and Albert Sauveur as secretary, made its final report. The definitions proposed, while doubtless as satisfactory as could be then formulated, left much to be desired from the standpoint of clearness and conciseness.

The definitions proposed by this Committee have been published so widely and are so well known that it is not necessary to reproduce them verbatim. All that has been done, therefore, is to indicate briefly how the various terms which denote the commercial iron-carbon alloys are used in this monograph and in the other monographs of the series. For this purpose the products of the iron and steel industry have been divided into six classes: (1) commercially pure iron, (2) wrought iron, (3) carbon steel, (4) cast and pig irons, (5) malleable iron, and (6) alloy steel.

1. Commercially Pure Iron, Wrought Iron, and Carbon Steel.—Commercially pure iron is, as the term implies, the element iron in as pure a form as can be, or as is, produced commercially and includes electrolytic iron, sponge iron, and open-hearth ingot iron. Data on these three classes, together with the available data on irons specially purified in the laboratory, form the subject matter of a previous monograph, "The Metal—Iron."[794] The most important of the commercially pure irons is, of course, open-hearth ingot iron which normally contains less than 0.03 per cent each of carbon, manganese, and sulphur, and even smaller amounts of phosphorus, silicon, and a number of other elements. According to an analysis given by Cleaves and Thompson[794] of a representative open-hearth ingot iron, the iron content by difference was 99.78 per cent. A few data on the properties of open-hearth ingot iron are given in the present monograph as typical of commercial iron containing traces to very small amounts of carbon.

Wrought iron is essentially a very low carbon steel, made by oxidizing, without completely melting, the charge and containing therefore a fairly pure ferrite matrix with which numerous slag

particles are intermingled. In hot working, these particles are elongated into stringers which give to wrought iron its well-known directional properties. Wrought iron usually contains less than 0.10 per cent carbon and manganese and less than 0.04 per cent sulphur. Phosphorus is ordinarily higher than in carbon steel, produced by melting and refining, but seldom higher than 0.20 per cent. The amount of wrought iron now being produced is an insignificant part of the world's total ferrous-metal production.

For the purpose of these monographs, carbon steel may be defined as a commercial iron-carbon alloy containing less than 1.7 per cent carbon (point E of the iron-carbon diagram) and varying but usually relatively small—less than 1 or 1.5 per cent—amounts of manganese and silicon and even smaller—less than 0.15 per cent—amounts of sulphur and phosphorus. In addition carbon steel contains traces and sometimes appreciable amounts of a large number of other elements, some of which form the so-called non-metallic impurities, including gases.

In a recent effort to formulate a definition of cast iron the British Cast Iron Research Association* proposed the following definition for steel. "*Steels:* Alloys of iron and carbon (other than malleable cast iron) with or without other elements, which do not contain carbide eutectic or graphite eutectic in the microstructure." This definition does not distinguish between ingot iron, which is usually an iron-carbon alloy, wrought iron, and low-carbon steel, by the slag content or by the refining process. It omits a specified carbon limit, usually placed at 1.7 per cent, because of the effect of some of the alloying elements on the location of point E.

The composition of carbon steel is discussed in greater detail below.

2. Cast Iron, Pig Iron, and Malleable Iron.—A wholly satisfactory definition of cast iron has never been formulated. The Committee on Uniform Nomenclature defined it as "iron containing so much carbon or its equivalent that it is not malleable at any temperature." The lower carbon limit for cast iron is nearly always placed at point E on the diagram, *i.e.*, at 1.7 per cent.

The recent definition by the British Cast Iron Research Association is as follows: "*Cast Irons:* Alloys of iron and carbon,

* NORBURY, A. L.: Cast Iron and Steel Differentiated, *Iron Age*, Nov. 7, 1935, pp. 25 and 104–112.

with or without the other elements, which contain carbide eutectic (white cast iron) or graphite eutectic (gray cast iron) or both carbide eutectic and graphite eutectic (mottled cast iron) in the microstructure." This is probably the most satisfactory definition yet proposed for cast iron. Pig iron, of course, is cast iron, cast directly into pigs from the blast furnace.

The definition, by the same body, for malleable iron follows directly from the definition of cast iron, *viz.*: "*Malleable Cast Iron:* The product obtained by eliminating the carbide eutectic from solid white cast iron by decarburizing (white heart), or by conversion into graphite by annealing (black heart)."

In the United States both the American Foundrymen's Association and the National Bureau of Standards are working on the problem of formulating an acceptable definition of cast iron.

3. Alloy Steel.—As yet no adequate definition of alloy steel has been formulated.* Although carbon and alloy steels cannot be separated with scientific precision, it is easily possible, at least for a number of the alloying elements, to determine practically whether the material falls within one or the other of these two classes. If, for example, an alloy, not added primarily for deoxidation, such as nickel, chromium, vanadium, molybdenum, copper, tungsten, etc., is intentionally added (either as the metal, as a ferroalloy, or as part of the charge) in the amount required to have the desired effect, the resulting product is clearly an alloy steel.

If, on the other hand, none of an alloy is intentionally added, or if a small amount is added primarily for deoxidation, the finished product is usually classed as a carbon steel even though a small adventitious amount of the element, resulting from the accidental admixture of alloy scrap or other material in the charge, may be found by analysis in the finished steel. This is only true, however, if the adventitious amount is so small that it does not have a marked effect (aside from deoxidation effects) on the structure or properties. Except for nickel, copper, and molybdenum, which do not oxidize and thus tend to increase in scrap over a period of years, most steels classed as carbon steel contain such a small amount of residual alloys that, for the present at least, these may be ignored.

* The American Iron and Steel Institute is working on this problem.

Most steel makers consider that, if steel contains less than 0.20 or 0.25 per cent nickel and less than 0.10 per cent copper or molybdenum, the material is carbon steel. Until further work is done on the effect of small amounts of alloys on the properties and structure of carbon steels (an important but woefully neglected field), these percentages, at least for the purposes of the monographs of this series, may be taken as the dividing line between carbon and alloy steels. It must be remembered, however, that these percentages may be used as a dividing line only in classifying steels in which nickel, copper, and molybdenum are determined by analysis. As they are not usually determined in routine steel-plant analysis of carbon steels, there is no way of telling how many steels have been made, rolled, and used as carbon steels which contained enough of one or more of these three elements to put them, by this arbitrary dividing line, into the alloy-steel class if the percentage of residual alloy were known.

In the case of the elements which oxidize in preference to iron, such as chromium, vanadium, tungsten, aluminum, titanium, and zirconium, there is usually such a small amount (less than 0.05 or 0.10 per cent at most) remaining after melting and refining that the heat can rarely be classed as an alloy steel. Specific data on the effects of small amounts of these alloys on the properties and structure of carbon steel are, except in the case of those elements which control grain size, almost wholly lacking; hence, until data are reported which show that less than 0.05 per cent, or 0.10 per cent in the case of chromium, of the element has a marked effect, it will be necessary to consider steels containing small residual percentages of these elements as carbon steels.

4. Silicon and Manganese in Carbon and Alloy Steels.—There is no way at present to differentiate between carbon steels containing considerable amounts of manganese and silicon, and manganese and silicon alloy steels containing low percentages of these elements.

Steels known throughout the industry as carbon steels may contain as much as 1.25 or even 1.50 per cent manganese and as much as 0.30 or even 0.50 per cent silicon; on the other hand, data are reported by Greiner, Marsh, and Stoughton(618) for "silicon-manganese" steels containing less than 1.00 per cent manganese and less than 0.50 per cent silicon. These authors quoted a specification, by the American Society for Testing Materials,

for "silicon structural steels" in which it is stated that the silicon shall be "not under 0.20 per cent," and in which the manganese content is not specified.

Regarding silicon, Camp and Francis[149] stated:

> Since silicon-treated steels may contain more than 0.50 per cent silicon and silicon alloy steels may contain as little as 0.50 per cent, the nearest custom comes to supplying a definition for silicon steels is to say that no steel containing only silicon, in addition to the carbon and manganese usually present in carbon steels, is to be considered as an alloy, or a special, steel unless the silicon content exceeds 0.50 per cent.

As few values for carbon steels, except in the case of cast material, reported in the following pages were obtained on material containing more than 0.30 per cent silicon, there is little, if any, duplication of data in the present monograph and in "The Alloys of Iron and Silicon."

Many of the carbon steels the data for which are reproduced and discussed in the following chapters contained around 1 per cent manganese. In correlating these data, an arbitrary dividing line of 1.25 per cent manganese was decided upon to separate carbon steels from low-manganese steels with, however, the full realization that a discussion of the properties of low-manganese steels in the forthcoming "The Alloys of Iron and Manganese" would include material designated as manganese steels and containing less than 1.25 per cent.

B. VARIABLES AFFECTING THE PROPERTIES OF COMMERCIAL IRON-CARBON ALLOYS

If commercial carbon steels were pure alloys of iron and carbon and if cast irons were pure alloys of iron, carbon, and silicon, there would be fewer problems involved in assembling and correlating data on their properties. What the result of such a correlation would be is an interesting speculation but quite beside the point, as there never have been, and most likely never will be, absolutely pure binary iron-carbon and ternary iron-carbon-silicon alloys made commercially.

Probably there are no commercial metallic materials which are equally hard to purify and in which small amounts of impurities have such a marked effect as in steel and cast iron. The amount and distribution of the impurities and the many other variables which may affect the constitution and properties of carbon steel

and unalloyed cast iron have been discussed in some detail in the first volume of this monograph (Chapters I, XI, and XII), hence only a brief summary is justified here.

5. The Composition of Carbon Steel.—The approximate amounts of the four elements which, in addition to carbon, are nearly always included in routine chemical analysis of carbon steels made by the principal steel-making processes are as follows:

Kind of steel	Approximate range, per cent			
	Mn	Si	S	P
Carbon steel	0.20 to 1.25	Trace to 0.50	0.01 to 0.08	0.01 to 0.12
Basic open-hearth tonnage steels, not deoxidized with silicon	0.30 to 0.60	Trace to 0.05	0.02 to 0.05	0.01 to 0.045
Basic open-hearth tonnage steels, deoxidized with silicon	0.50 to 1.25	0.10 to 0.30	0.02 to 0.05	0.01 to 0.04
Acid Bessemer tonnage steels, mostly low-carbon	0.30 to 0.90	Trace to 0.02	0.02 to 0.065	0.07 to 0.12

More than 98 per cent of the carbon steel produced in this country falls into the last three classes given above. All of the above four elements affect the properties; manganese and phosphorus to a greater extent (at least for the amounts usually present in carbon steels) than silicon and sulphur. In basic open-hearth carbon steels the manganese may vary widely and, consequently, be mainly responsible for the marked variation in properties frequently noted in basic steels of the same carbon content. The much greater percentage of phosphorus in acid steels than in the basic product has a profound effect on the mechanical properties of material otherwise of the same composition.

Not only does the composition of steels of the same carbon content, as ordinarily determined, vary within fairly wide limits from grade to grade and from heat to heat, but, because of segregation, it may also vary from ingot to ingot, billet to billet, and even from bar to bar in the same heat.

The first two factors, therefore, which may affect the properties of commercial iron-carbon alloys of the same carbon content are: (1) the amount of manganese, silicon, sulphur, and phosphorus in steels of different grades, or made by different processes, or in different heats; and (2) the amount of these elements when segregated in steel of the same grade, made by the same process, and in the same heat.

In addition to accidental, occasionally large but usually small, amounts of alloying elements such as nickel, chromium, vanadium, copper, molybdenum, and others which may be present, commercial carbon steels may also contain small amounts of other elements, usually picked up from the charge. Among these, mentioned by Epstein in Volume I of this monograph, are arsenic, tin, lead, zinc, antimony, and aluminum and occasionally titanium or zirconium from the deoxidizers used. Except for the control of grain size by aluminum, the effect of such accidental small amounts of these elements is not known. Furthermore, these impurities are rarely determined, so that there is no information available on how frequently traces of these impurities may be encountered in steel or cast iron.

Also present, but determined infrequently, may be traces or even appreciable amounts of non-metallic material such as sulphides and arsenides, of gases, especially oxygen, nitrogen, and hydrogen, and their reaction products. That oxygen, especially in the form of oxides and silicates, and nitrogen affect the structure and properties is certain; little or nothing is known about hydrogen except that it apparently causes acid brittleness in pickling.

The next two factors, therefore, which may influence the properties of commercial iron-carbon alloys of the same carbon content are: (3) the accidental, small, and usually undetermined amounts of metallic impurities which may be picked up from the charge; and (4) the variable amounts—usually small but sometimes appreciable—of gases and their reaction products and other visible or invisible solid non-metallic inclusions which may be present. This last factor has an especially important effect on the directional properties, on the resistance to single-blow impact, and in some instances on the resistance of a steel to repeated stresses.

6. Other Variables Which May Affect the Properties of Carbon Steels.—There are a number of factors, in addition to the composition, which may affect the properties of iron-carbon alloys. A large part of these may be brought together under the broad heading of variations in microstructure. It is hardly necessary to do more than mention that the properties of a 0.50 per cent carbon steel vary with the relative amounts of ferrite, lamellar pearlite, spheroidized cementite, troostite, martensite, or austenite present. The most important variable, in addition to the composition, which affects the microstructure is, of course, the thermal treatment. Others are the mechanical treatment and the size of the section. The size, number, or distribution of visible and invisible inclusions are also complicating variables in determining the final microstructure of the test specimen.

Another factor, related both to the composition and to the kind, number, size, and distribution of the inclusions, is the effect of the steel-making process; namely, whether the steel was made by an acid or a basic process, under strongly oxidizing, mildly oxidizing, or reducing conditions, how it was poured into molds, and how rapidly it solidified after pouring. Although the character of the steel-making process may be considered as a separate factor, it is so thoroughly entangled with differences in composition and with the amount and character of the visible and invisible inclusions that no one can say definitely whether a steel of a certain carbon content, melted and refined in a basic open-hearth furnace, would differ in properties from a steel of the same carbon content made in an acid Bessemer converter, if these other variables were not complicating factors.

The last factor, which has been receiving much animated attention lately, is the tendency in some steels, melted and refined in a certain way, for grain growth to occur at much lower temperatures than in other steels of practically the same composition, thus producing large differences in grain size in similar steels, similarly treated, and the tendency of some steels to age harden. According to Epstein (Volume I, Chapter XII, of this monograph), these two tendencies are the components comprised in the old term "body," which was once widely used to describe the difference in resistance to dynamic stresses and the marked differences in response to heat treatment in two steels of the same composition as determined by the usual analysis.

The three other factors, then, which may affect the properties of steels of the same carbon content are (5) the effect of the microstructure, which depends upon the composition, thermal treatment, cross-section, and mechanical treatment; (6) the effect of the melting, refining, and pouring practice; and (7) the effect of certain tendencies—the causes of which are mostly obscure and the terminology of which is not yet established—for grain growth and age hardening under certain conditions.

There are almost as many variables affecting the properties of cast iron as there are influencing the properties of the lower carbon alloys. First and most important, of course, is the composition. Others are the conditions of melting—including the nature of the process, the melting or superheating temperatures, and conditions of solidification and cooling—size of the section, thermal treatment, and the so-called heredity effect which includes the effect of visible and invisible inclusions.

7. The Accuracy of Available Mechanical-property Data on Carbon Steels.—There are probably no published data on the mechanical properties of carbon steels in the determination of which all of the above variables are considered. In the first place, the determination of the effect of some of them is difficult or even impossible at present because methods are imperfect or not available at all. In the second place, investigators do not usually take into consideration all of the factors which may affect the results, even when methods are available, either because it would complicate their problem too much or because they do not recognize that these variables are important. In the third place, investigators are often careless about reporting all of the facts which are known. Thus a paper on the properties of carbon steel may contain the analysis for manganese, silicon, sulphur, and phosphorus, and even for some of the more common alloying metals; it may, and usually does, give the thermal treatment (usually however with many important details lacking); it may give the size of the section treated, and may even give details of the mechanical working and the process by which the steel was made, but it usually neglects to give any details of melting and refining, pouring practice, location of the specimens with respect to a specific part of the ingot, or temperatures of rolling, and the like. There is almost never any information on adventitious impurities or on the contained gas and non-metallic

inclusions even as determined by the imperfect methods now available. Until very recently, the effect of grain-growth tendencies was unknown and, of course, ignored.

It follows from this that few, if any, of the properties given in the following pages should be considered as final or complete. Fortunately, however, in much of the published work the effect of only one variable was investigated and the other variables were, perhaps unwittingly, kept constant. For example, if an investigator determined the effect of tempering temperature on the tensile properties, it is usually safe to assume that his specimens were all from the same heat and probably from one single bar in the heat. This would, of course, reduce greatly the chances for large variations in composition, in melting, refining, pouring, and rolling practice, in adventitious impurities, in non-metallic inclusions and gas, in grain size, and in segregation.

Thus, the properties correlated in the following pages should show trends due to certain variables with fair accuracy. Most of the data reported in this volume are of this type. Comparisons of the data of one investigator with those of another—a few such comparisons are made—are of exceedingly questionable reliability. General conclusions on the effect of such broad variables as the melting practice and a few others cannot be drawn with any degree of accuracy from the data available; statistical analysis should assist in making such conclusions more reliable.

C. TENSILE TESTS

One of the most important factors for the reliability and reproducibility of mechanical-property data is the method of testing. Although this and other monographs of the series are not concerned with testing *per se*, some mention of testing methods is justified, especially as the interpretation of the data presented in these monographs, and the comparison of these with other data, cannot be made readily unless some details of the testing methods and their effect on the results are known.

Testing methods may be divided broadly into two classes: (*a*) standardized and (*b*) unstandardized tests. In the case of the former, it usually is not necessary to report all of the testing details; with unstandardized tests, however, full details of the method, including a description of the specimen used, are important and should always be reported together with the data

obtained. Unfortunately in many of the reports in the literature, even in some of the most comprehensive ones, this is not done. In this event a reviewer of the literature has the alternative of ignoring such reports, or of accepting the values reported as comparable among each other on the assumption that the original investigator at least maintained uniformity of method and specimens throughout his investigation.

After careful consideration it was decided to devote some space in the later chapters of this monograph to methods of making such unstandardized tests as those for hardness, impact, endurance (which is now standardized to a greater degree than some of the others), creep, damping, corrosion, deep drawing, machinability, and wear. In some of these—for example, machinability and wear, where the data obtained depend solely upon controlling closely every testing condition and are valid only for those conditions—the very nature of the subject precluded any more than a brief discussion of the tests and their practical value, if any; in other cases more details could be given.

8. Methods Used for the Determination of Tensile Properties. The four static properties (tensile strength, yield strength, elongation, and reduction of area) are determined by methods which are fairly well standardized. Although there are many variables in making these tests which need to be defined more carefully, the uniformity in the determination of these four properties, throughout the world, is sufficient—at least in comparison with such tests as those for machinability, wear resistance, and the like—so that results obtained in one country are comparable with fair accuracy with similar results obtained on like material in another country.

Much valuable work has been done on the standardization of tensile testing in this country by the Committee on Mechanical Testing of the American Society for Testing Materials.* Particular attention has been paid to the dimensions of specimens, to the speed of testing, and to the determination of the property loosely termed, especially in the older literature, elastic limit, proportional limit, or yield point.

* The following specifications have been issued: E8-33; E6-32T; E4-33T; E9-33T. These have recently been reviewed by R. L. Templin, chairman of the Section on Tension Testing, in *Metal Progress*, February, 1935, pp. 29–32.

Tensile Strength.—Although the absolute accuracy of tensile strength depends upon the dimensions of the specimen, the speed of testing, the sensitivity of the machine, and the alignment of the specimen in the machine, the values for this property as reported throughout the world probably are as accurate and as comparable as the values for any mechanical property. It is not possible to give the probable amount of error in the tensile-strength values reported in this and the other monographs of the series; it may, however, be assumed that, when the tensile strength of a steel is reported as 33.5 tons per sq. in., 52.7 kg. per sq. mm., or 75,000 lb. per sq. in., it has an actual tensile strength not far removed from these values.

Yield Strength.—The yield strength has been defined by the American Society for Testing Materials as the stress at which a material exhibits a specified limiting permanent set. This is measured by determining from the stress-strain diagram the stress necessary to produce a set of 0.2 per cent; or, for ductile materials, by the drop of the beam of the testing machine or by a pair of dividers. The methods for the determination of yield strength are standardized and are reproduced in practically all metal specifications and handbooks of testing.

Yield strength, determined in routine testing by the drop of the beam or by dividers, was formerly called yield point; this term is still used widely in the United States and England. The German property "Streckgrenze," always here translated as yield strength, is the drop-of-beam yield point or the stress producing a set of 0.2 per cent.

The only other value similar in character to yield strength used in this series of monographs is proportional limit, which is defined as the greatest stress which a material is capable of developing without deviation from Hooke's law. As the proportional limit, when determined carefully from the stress-strain curve, is for all practical purposes the same as the elastic limit, the latter term is not used at all. Parenthetically it may be said here that the values for elastic limit in the older reports, especially as determined in routine testing laboratories, are actually drop-of-beam yield points and so are always reported as yield strength. Practically, the proportional limit, as used in these monographs, is a special yield-strength value in which the limiting set is not measurable with the instruments used and, hence, is considered

to be zero. In a few proportional-limit values in the following chapters the limiting set was 0.05 per cent.

The yield-strength values reported in these monographs may or may not be comparable. In most German reports the limiting set for this value is known; in the other reports it usually is not known and so may not be comparable with values given in German reports. In general, however, the values in any one report are comparable with each other.

Elongation.—Elongation values are directly comparable only when the cross-sectional area of the specimen bears a fixed relation to the gage length. In the United States the gage length of round specimens which differ from the standard 0.505 × 2-in. test bar is calculated by the formula

$$G = 4.5\sqrt{A}$$

where G is the gage length and A the cross-sectional area in square inches. So long as all specimens conform to this relationship, elongation values are comparable.

Values determined on U. S. A. standard round specimens (0.505 × 2 in.) which have a cross-sectional area of 0.2 sq. in. are, however, not comparable with the values obtained on standard British specimens having a diameter of 0.564 in., a cross-sectional area of 0.25 sq. in., and a gage length of 2 in.

Two types of round specimens are used in Germany, the long specimen, with a diameter of 20 mm. (0.787 in.) and a gage length of 200 mm. (7.875 in.), and the short specimen with a diameter of 20 mm. (0.787 in.) and a gage length of 100 mm. (3.937 in.). The elongation values obtained with these specimens are not directly comparable with British or American values as the formulas for the German specimens are

Long:

$$G = 11.3\sqrt{A}$$

Short:

$$G = 5.65\sqrt{A}$$

where A is given in square millimeters.

The standard round specimens used in France have a diameter of 13.8 mm. and a gage length of 100 mm. The formula is then

$$G = 8\sqrt{A}$$

All the elongation values reported in this series of monographs were, unless otherwise specifically stated in the text, obtained on U. S., British, German, or French standard round specimens of the dimensions given on the preceding page.*

Reduction of Area.—As values for the reduction of area are expressed as a percentage of the original area, all such values obtained on standard round tensile specimens should be roughly comparable.

D. STATISTICAL METHODS APPLIED TO MECHANICAL PROPERTIES

For many years workers in such diverse fields as physics, biology, sociology, insurance, business administration, and many others have made use of statistical methods to analyze the large volume of numerical data accumulated in these fields. The use of statistical analysis has many advantages. It largely eliminates the personal factor in the interpretation of test data; it supplies important information which can be obtained only with great difficulty, if at all, by other means; it frequently shortens greatly the time necessary to interpret correctly a large volume of data, and it gives important information on the character of the variables present, whether they are or are not beyond the control of the investigator.

9. Importance of Statistical Analysis.—The importance of statistical analysis for the ferrous and non-ferrous metal industries has been realized only in the last few years. This is shown by the growing literature on the subject, mostly published since 1930. It is out of place here to review this literature or to give a comprehensive bibliography, but the work of Daeves at the Vereinigte Stahlwerke in Germany, of Shewart at the Bell

* For a more detailed discussion of elongation than is given here the reader is referred to C. A. Bertella: Sulle prove di tensione dei metalli. Relazione fra l'allungamento di rottura e le proporzioni delle barrette (Tensile Tests of Metals. Relation between Elongation and the Proportions of the Test Specimens), *Giornale del Genio Civile*, v. 60, 1922, pp. 343–368, who in 1922 developed a formula which expresses the precise fundamental relationship between cross-sectional area, gage length, and elongation. This formula has been discussed by R. L. Templin: Effects of Size and Shape of Test Specimen on the Tensile Properties of Sheet Metals, *Proc. Am. Soc. Test. Mat.*, v. 26, part 2, 1926, pp. 378–402, and by D. A. Oliver: Proposed New Criteria of Ductility from a New Law Connecting the Percentage Elongation with Size of Test Piece, *Proc. Inst. Mech. Eng.*, 1928, pp. 827–864.

Telephone Laboratories, New York, of Committee E-1 of the American Society for Testing Materials, and of the British Standards Institution should be mentioned.*

Earlier in this chapter it was emphasized that practically every factor in the production of a finished steel section is subject to variations and that these variations affect the properties. There are, perhaps, a few properties of iron-carbon alloys which can be stated in absolute terms—specific gravity, heat of fusion, thermal conductivity, and the like—but only a few; in the case of most of the important engineering properties the best that can be done for any specific class of commercial iron-carbon alloy is to give an approximation. When a steel maker says that he will supply deep-drawing sheet with a tensile strength of 50,000 to 55,000 lb. per sq. in. he means that the material he produces with his standardized practice has a strength within this range most of the time; but he knows that there will be a relatively large number of heats, and occasionally a few sheets of the same heat, which will have strengths above or below the required limits. This is true, not only of tensile strength, but also of all the other important engineering properties.

* K. Daeves has published a large number of papers in the German technical press and has summarized much of his work in his book, "Praktische Grosszahl-Forschung" (Practical Statistical Investigations), VDI Verlag, Berlin, 1933, 132 pp. W. E. Shewart is author of a large number of papers, and his book, "Economic Control of Quality of Manufactured Product," D. Van Nostrand Co., Inc., New York, 1931, 501 pp., is a broad and comprehensive discussion of the whole subject. The "Manual on Presentation of Data" published in August, 1933, and the supplements to this manual issued later in the same year by the American Society for Testing Materials give methods of applying statistical analysis to the interpretation of test data. Other papers which should be mentioned are the elementary discussion by R. F. Ferguson on "The Interpretation of Plant and Laboratory Test Data," in *J. Am. Ceramic Soc.*, v. 13, 1930, pp. 354–362, and the paper (of great practical interest to steel men) by W. C. Chancellor on the "Application of Statistical Methods to Metallurgical Problems in the Steel Plant," in *Proc. Am. Soc. Test. Mat.*, v. 34, part 2, 1934, pp. 891–919. Just published by the British Standards Institution is a 160-page booklet on "The Application of Statistical Methods to Industrial Standardization and Quality Control" (No. 600, November, 1935). A well-known standard work is R. A. Fisher's "Statistical Methods for Research Workers," Oliver and Boyd, London, 5th ed., 1934, 333 pp. All these publications contain references to earlier and in some cases to mathematical dissertations on statistical theory.

Obviously then, it should be of vital importance to be able to tell with accuracy whether this variation in properties represents a normal variation resulting from chance and is to be expected under given controlled conditions of operation, or whether it is due to assignable causes which can be traced to their source and thus perhaps placed under control. It is the object of statistical analysis to tell this.

According to Chancellor*

it is undoubtedly true that *variability* in connection with steel making and *relationship* between steel-making variables constitute major problems in this field. [This is as true for variables which affect mechanical properties as it is for the variables which affect melting and refining.] There is a general tendency to deal with averages in speaking of steel-plant problems, and in spite of the fact that average values are very often satisfactory, the variation over and under the average constitutes the real source of the difficulty. A knowledge of the degree of variability and some conception of the causes of such variability will help materially in defining and classifying steel-works problems.

10. Statistical Methods Applied to Mechanical Properties.— It is impossible, in the space available in this chapter, even to touch upon the fundamentals of statistical analysis. This is not necessary, however, as these methods are clearly outlined in nearly all the works cited in the footnote on page 17. In brief, statistical analysis consists in arranging the data (for example, tensile-strength values) in numerical order and determining the arithmetic mean and the standard deviation from this mean. (The standard deviation is the square root of the mean of the squares of the deviations of all of the observations.) When only a few observations are available, these are usually plotted in a dot diagram; when a larger number of results are at hand, these are divided into groups, usually 15 to 25 in all, and the number of observations falling in each group is counted up. This number is the frequency and is used in plotting the familiar bell-shaped frequency diagram. (See Fig. 4, page 44.)

In the usual frequency diagram most of the values are clustered around the central value in the upper part of the bell, and the values which differ greatly from this central value are very much fewer in number. This distribution is so nearly universal

* *Op. cit.*, see footnote p. 17.

that it is called the normal distribution. It has been found that the standard deviation from the arithmetic mean has the same significance for all normal distributions. Only one result in three will have a deviation greater than the standard, thus 67 per cent of all the values will be within the upper part of the bell of the frequency diagram and will include the mean plus or minus the standard deviation. Deviations greater than twice the standard will occur in one out of each 23 values, and deviations greater than three times the standard will occur once in 370 specimens.

The frequency curve and the standard deviation can be used to disclose many important facts which would be overlooked in the ordinary routine examination of a large number of test results and a simple average value of these results. A single example will illustrate this. Suppose that over a period of a week low-carbon sheet made by the regular practice had an average tensile strength of 50,000 lb. per sq. in. and a standard deviation of 1500 lb. per sq. in., or 3 per cent. Suppose further that the sheet made the following week, apparently by the same practice, had an average tensile strength of 50,000 lb. per sq. in., but that in this material the standard deviation was 1800 lb. per sq. in., or 3.6 per cent. A cursory examination of the individual test results would probably show no more "freaks" than in the previous week's production, and the average tensile strength of all the specimens is the same. The change in standard deviation has, however, indicated that somewhere between melting and testing the finished sheet the practice has actually changed.

Chancellor summed up the question of the use of statistical methods in the steel plant by saying:

> The advantage in using a statistical analysis lies in the more exact formulation of the conditions to be corrected, since detectable and non-detectable causes of variation are differentiated and efforts can be efficiently directed either toward better control of the old process or adoption of a more satisfactory one, as indicated by the results of such statistical analysis.

The value of statistical methods in analyzing steel-works data has frequently been emphasized by Daeves;* some of Daeves' actual data are quoted in Chapter IV.

* *Op. cit.*, see footnote p. 17.

CHAPTER II

VARIABLES AFFECTING THE PROPERTIES OF CAST STEELS

Steel Castings as Industrial Products—Testing Methods and Heat Treatment—Solidification, Recrystallization, and Shrinkage—Effect of Chemical Composition on Properties—Effect of Size of Casting and Location of Test Specimen on Mechanical Properties—Author's Summary

The steel-melting furnaces of the United States produce, in a normal year, between 40 and 50 million tons of molten metal all of which is teemed into molds and solidifies as cast steel. Between 97 and 98 per cent of this production is cast into ingots, which are processed further into a large variety of hot- and cold-worked products. The remainder, varying usually from 800,000 to 1,500,000 tons, is poured into sand or metal molds which are of the shape desired for the finished product. When the casting has solidified, it is placed into use without any mechanical treatment.

Although the properties of cast steel are generally considered to be inferior to the properties of a mechanically worked material of similar composition, the ratio of production of steel castings to hot- and cold-worked products has remained substantially constant for many years. There are two reasons for this. First and most important: steel may be cast into sections of such intricate shape that their production by rolling or forging is either impossible or prohibitively expensive. Second: for the manufacture of a limited number of pieces of irregular cross-section, castings are cheaper than mechanically worked material of the same size and shape. It is only by a large production of the same piece that the high cost of forging dies can be distributed so that forgings can compete in price with castings.

Cast steel is of interest to the metallurgist, steel worker, and steel user for two equally important reasons: (1) it is a valuable industrial finished material and (2) it is the intermediate product for all wrought steels and, consequently, variations in its

structure and properties may be reflected in the structure and properties of the hot- or cold-worked sections made from it.

Before reviewing the mechanical properties of cast carbon steel, it is advisable to consider in this chapter the method of taking test specimens and the heat treatment of castings, as well as the factors which may affect the quality of the castings and thus directly or indirectly affect the properties.

To prevent ambiguity in the following discussion, the terms *cast steel* and *steel casting* are defined as follows:

Cast steel, a commercial iron-carbon alloy containing less than 1.7 per cent carbon, melted, and poured into a mold of any material, size, or shape. Cast steel may be used commercially in the form in which it is cast or it may be worked hot or cold into an industrial product.

Steel casting, cast steel which has been poured into a mold of the shape desired for the finished product and is so used industrially without any mechanical treatment.

A. STEEL CASTINGS AS INDUSTRIAL PRODUCTS

According to Hamilton,[(526)] the production of steel castings in the United States began about 1867. In early production, dry-sand molds were used; green-sand molds were not developed until about 20 years later. The peak production of steel castings was 1,585,000 tons in 1929.*

11. Manufacture and Use of Steel Castings.—The various steel-making processes were described briefly in Chapter I of the first volume of this monograph. In general, the same processes are used for melting steel for castings as for ingots, but the furnaces, especially in the case of the open-hearth, are smaller. The production of castings distinguished by processes varies slightly from year to year but is distributed chiefly between the acid and basic open-hearth and the electric process, as the percentages for two typical years indicate (see tabulation on page 22).

The bulk of the tonnage is produced in the open-hearth furnace, although the electric-arc furnace is becoming a more important factor. The increase in use of electric-arc furnaces, especially those with an acid lining, may be attributed to their flexibility, particularly in temperature control, and to some extent to the relative ease with which the furnace atmosphere can be varied.

* Annual Statistical Report of American Iron and Steel Institute, 1931.

Process	Percentage of total production	
	1928	1933
Basic open-hearth	31.4	25.0
Acid open-hearth	37.6	38.2
Electric	28.6	35.8
Bessemer	2.3	0.9
Crucible	0.1	0.1

One of their most important advantages is their small size which readily permits economical intermittent operation. This advantage is especially marked during periods when demand for castings is relatively small.

High-frequency induction furnaces, which have become important in melting steel for high-grade alloy- and tool-steel ingots, are finding use in the production of high-alloy castings, especially those of low carbon content. This has recently been described by Middleton.(553)

Most of the electric furnaces used for melting steel for castings have an acid lining. The choice of a lining for open-hearth furnaces, especially those used for large castings, depends chiefly upon the preference of the furnace owner, and upon the location of a plant with reference to sources of raw materials; an acid lining is ordinarily used when an adequate supply of high-grade low-phosphorus scrap is available, and basic when the melting stock requires considerable refining.

In the introduction to this chapter it was mentioned that steel castings can be used advantageously when their intricate shape makes the use of forged or rolled sections impossible or uneconomical, or when there are so few pieces needed that the high unit cost of forging dies is prohibitive. In addition, there are some industrial applications for carbon steel where bulk or weight is more important than strength, so that the lower mechanical properties of the cast material are compensated for by the greater cross-section. When these conditions prevail, castings are used because they are cheaper.

Castings are used in a great variety of industries and in a wide range of weights and sizes, from a few ounces to 200 tons or

over, and from a few inches in the maximum dimension to engine beds 65 ft. or more in length. While the railroads have been the largest users of steel castings, at the present time sizable tonnages are also used for agricultural, excavating, and oil-well machinery, road-building and steel-making machinery, electrical machinery, and machine tools.

12. Necessity for the Heat Treatment of Steel Castings.—In Chapter VII of the first volume of this monograph it was stated: "the Widmanstätten structure [coarse, angular structure found in castings] is associated with brittleness even in steel castings of low carbon content, and one of the primary reasons for the annealing or normalizing of steel castings is to eliminate this structure."

The date of the first use of heat treatment is uncertain. According to Hall,(256) the Taylor-Wharton Iron and Steel Company marketed in 1909 the first heat-treated carbon-steel castings manufactured in America. Hamilton,* however, and Bull* stated that in 1898 a Chicago company was heat treating cast-steel car wheels.

When heat treatment was first used, the castings were simply annealed† or normalized, but this was soon followed by more complicated treatments, such as normalizing followed by annealing, normalizing followed by tempering at a high temperature (usually just below the critical range), a double normalizing followed by tempering, and others. Quenching in a liquid medium has also come into extensive use, probably due to the increased introduction of alloy castings which require rapid cooling to develop their potential properties. Lorenz(543) stated that quenching of alloy-steel castings in water or oil is capable of developing tensile- and yield-strength characteristics at least 50 per cent higher than those typical of the metal in the normalized condition, without appreciable loss in ductility.

Lorenz also stated that in his experience with many thousands of tons of steel castings which had been given a water-quenching treatment the scrap was only a few hundredths of 1 per cent and "these cases could invariably be traced to prior causes." While

* Private communication.

† Throughout this volume the term "anneal" is used to designate the operation of heating to above the critical temperature and cooling slowly. This treatment is the one designated in A.S.T.M. specification A119-33 as "full anneal."

it is well known that benefits from a liquid quench of alloy-steel castings are greater than in the case of carbon-steel castings, Lorenz reported that "liquid-quenched low-carbon steel castings are being produced in considerable tonnages at the present time (1932)." Mitchell[393] also emphasized the importance of the water-quenching treatment for medium-carbon steel castings and presented the supporting data shown on page 65 and in Fig. 7, Chapter III. Gregg[617] reported interesting results obtained by water quenching locomotive-wheel centers; this is also discussed in the next chapter.

The use of high temperatures in tempering normalized or quenched steel castings as a means of improving the ductility has been advocated by MacPherran and Harper,[129] Harper and Stein,[122] Lorenz,[268, 543] and others. MacPherran and Harper showed that by tempering as high as 700°C. (1290°F.) good ductility could be obtained with specimens drilled from castings which weighed up to 100,000 lb. The improved ductility was attributed largely to spheroidization of the cementite. Similar conclusions were reached by Harper and Stein, Lorenz, and others. In some cases the castings have been heated to just above the lower critical point as a means of hastening the spheroidization of the cementite.

13. Specifications for Steel Castings.—Current specifications for carbon-steel castings have been issued by the American Society for Testing Materials, the Society of Automotive Engineers, the U. S. Government (Master Specifications), U. S. Army, U. S. Navy, and the American Railway Association. Most of these are collected in the 1933 volume of specifications issued by the National Bureau of Standards.[644] The American Society for Testing Materials, the U. S. Government Master, and the U. S. Army Specifications give the following minimum mechanical properties for the three common grades of castings:

Grade	Tensile strength, lb. per sq. in.	Yield strength, lb. per sq. in.	Elongation in 2 in., per cent	Reduction of area, per cent
Hard	80,000	36,000	17	25
Medium	70,000	31,500	20	30
Soft	60,000	27,000	24	35

The Society of Automotive Engineers specifies the same tensile strength and yield strength, but the elongation for the three grades is 15, 18, and 22 per cent, and the reduction of area, 20, 25, and 30 per cent. Some specifications have a minimum cold-bend value of 90 deg. for the medium and 120 deg. for the soft grade. Some specifications give minimum mechanical properties for special castings, in addition to those for the three common grades, although it is the general practice to leave these unspecified.

Most specifications give the maximum permissible content of sulphur and phosphorus as 0.06 and 0.05 per cent respectively. For acid steel a higher maximum phosphorus content, 0.06 or 0.07 per cent, is necessary. The maximum manganese, when specified at all, is 0.80 or 0.85 per cent with 0.50 per cent minimum specified in one case. Only two specifications give silicon contents; these are 0.20 per cent minimum in one and 0.40 per cent minimum in the other.

As minimum mechanical-property values are given in most of the specifications, the carbon content is usually not included. The customary percentage ranges from 0.50 per cent for hard castings to 0.15 per cent for soft castings. Formerly, manganese and silicon as well as carbon percentages were not given. The inclusion of these two elements in some recent specifications was commended by Bull[(442)] as a means of making the production of sound castings more easily possible. All authorities however are not in agreement on this point. For example Merten[(273)] concluded that silicon should not exceed 0.20 per cent in large castings. Bull maintained that carbon, manganese, and silicon must be considered collectively, and that foundrymen should be free to adjust these elements to meet the physical requirements specified by the purchaser.

B. TESTING METHODS AND HEAT TREATMENT

The method of casting the test specimen is a subject which has received much attention and some experimental study. This problem does not appear to be serious in the case of the smaller castings but becomes of increased importance in larger castings. Foundrymen, metallurgists, and testing engineers are agreed that small specimens attached to large castings or separate blocks are not representative of the large castings in either structure or properties as cast and cannot be made so by the same heat treat-

ment. The problems of gating and feeding the test specimen when attached to the casting must be given consideration, so as to produce a sound test specimen and at the same time avoid inducing unsoundness in the casting to which it is attached. Since longer holding times at the heat-treating temperatures are required as the size of the casting increases, and higher temperatures are also generally employed, it is evident that either the specimen might be overheated or the large casting might be underheated when they are given the same heat treatment. The method of casting the test specimens was discussed at length by Rosenhain(561) and others without reaching complete agreement.

14. Test Specimens and Method of Testing.—The test specimens are generally required to be attached to the casting or a block representative of the casting and are not to be removed until the heat treatment is completed. In small and therefore inexpensive castings test specimens may be cut from the casting when this is satisfactory to the manufacturer. The tensile-test specimen is 0.505 in. in diameter and has a 2-in. straight gage length with ⅛-in. radius at the shoulder. The bend-test specimen is 6 in. long and has a cross-section of 1×0.5 in. with the corners rounded to a radius of not over 1/16 in.; it is required to bend around a pin 1 in. in diameter by the amounts specified without fracturing the surface in tension.

The yield strength is usually determined by the drop of the beam of the testing machine. The maximum speeds of the crosshead of the testing machine are usually specified, the common values in inches per minute for the yield strength and the tensile strength being ⅛ and 1.5 respectively.

15. Variable Methods of Taking Specimens.—In an effort to solve the problem of the difference in structure and properties of large castings and of the small test coupons attached, Merten(273) and Harper and Stein(122) resorted to the use of core drills to take test specimens from the heavy sections of castings or from comparable sections of test blocks as a means of determining the properties of the metal in such castings. Obviously, all large castings cannot be cored for test specimens, and the problem then resolves itself into two possible procedures: (1) to cast test blocks with sections comparable to the heaviest section of the casting, and examine specimens from different parts in such a way

that an estimation of the properties of the casting can be made; (2) to cast a standard-size test block with the specimens attached, and from the properties found in these specimens and proper size factors estimate the mechanical properties of the casting. The latter method requires adequate knowledge of the effect of size and of the necessary heat treatments to produce certain probable properties in large castings. In each procedure as much consideration must be given to design, gates, risers, etc., of the test block as is required for sound castings.

Some interesting experiments by Heinrich on specimens cut from axle housings and on separately cast test castings are discussed on pages 43 and 44, where other data on the relation of size of section to the properties and to the required heat treatments are given.

16. Heat Treatment.—Some of the standard specifications for steel castings require that the castings be heat treated, others do not. The following heat treatments are recognized by the American Society for Testing Materials:* (*a*) annealing, (*b*) normalizing, (*c*) normalizing followed by reheating, and (*d*) quenching followed by reheating. The American Society for Metals† included another heat treatment, "double normalizing and annealing," in which the castings are first normalized at relatively high temperatures, 870 to 1095°C. (1600 to 2000°F.), for the purpose of homogenizing the structure, after which they

TABLE 1.—RECOMMENDED ANNEALING AND NORMALIZING TEMPERATURES FOR CARBON-STEEL CASTINGS

A.S.T.M.*			A.S.M.†		
Carbon, per cent	Temperature		Carbon, per cent	Temperature	
	°C.	°F.		°C.	°F.
Up to 0.40	900	1650	0.15 to 0.40	870 to 930	1600 to 1700
0.40 to 0.60	850	1560	0.40 to 0.60	845 to 870	1550 to 1600
Over 0.60	830	1525			

* Standards, 1930, part 1, p. 417.
† "Metals Handbook," 1936, pp. 735–737.

are annealed at only slightly above the critical temperature to refine the grain.

Recommended temperatures for single annealing or normalizing castings of different carbon contents are given in Table 1.

It is generally recommended that castings be held (soaked) at the above temperatures 60 min. for each inch in maximum section to provide for thorough heating and completion of the desired structural changes. Only two specifications (A.S.T.M. A27-24 and A95-29) definitely prohibit quenching the castings in a liquid medium.

17. Radiographic Examination.—While none of the specifications for steel castings includes radiographic examination, it is well known that castings for special types of service are being purchased subject to satisfactory results from radiographic tests of each individual casting or of a certain number of representative castings from a lot. Both X-rays and gamma rays are being used. Radiographic testing has served as a means of inspection for acceptance and as a research tool in developing sound castings. The applications of radiographic methods have been discussed by numerous authors; *e.g.*, Lester,[128] Pullin,[280, 404, 480, 558] St. John,[336] Briggs and Gezelius,[441, 594] and Mehl.[549]

Isenburger[804] stated that when the same thicknesses of cast steel and wrought steel are X-rayed, an exposure that is correct for wrought steel will be too long for cast steel and ascribed this to a difference in density. He found this particularly evident when the material is over 2 in. thick.

C. SOLIDIFICATION, RECRYSTALLIZATION, AND SHRINKAGE OF STEEL CASTINGS

Because of their importance in steel castings, some of the phenomena observed in the solidification, cooling, and heating of iron-carbon alloys require special consideration. The hot working to which wrought alloys are subjected eliminates, or at least tends to eliminate, the cast structure in such products and at the same time helps to modify the secondary structures. As a result, the wrought steels are usually more responsive to the various heat-treating operations.

18. Solidification.—The solidification of iron-carbon alloys has been discussed in Volume I, Chapters II and XI, of this

monograph. In steel castings containing less than about 0.30 per cent carbon the first product to solidify is delta iron, which reacts peritectically with the liquid phase at about 1490°C. (2715°F.). For the lower carbon alloys (up to about 0.20 per cent carbon), this reaction results in the completion of the solidification. In the higher carbon alloys, the delta phase disappears and leaves gamma iron in equilibrium with the molten metal. The influence of the peritectic reaction on the properties of steel castings is not known. It is known, however, that in the solidification of steel castings characteristic structures are produced which are referred to as casting structure, ingotism, and the like.*

Solidification begins at the mold surface and progresses inward; thus crystals form at the surface of the casting and grow inward. The size of these crystals increases with the size of the casting and with decreased rates of cooling. The first crystals to form are relatively pure iron, and the last material to solidify is richer in carbon and impurities. In castings, the growing crystals may take the form of columnar crystals or dendrites, leaving the surrounding or filling-in material high in impurities. Evidently, this produces a segregated structure which usually affects the properties of the steel castings when they are tested in the cast condition and is shown by relatively low ductility. One of the important functions of heat treating steel castings is to homogenize this segregated structure. Since the size of the crystals increases with increase in the solidification interval and, therefore, with increased size of castings, it follows that longer periods of time must be provided in heat treating large castings than is necessary for small castings. High heat-treating temperatures also aid in diffusion and, therefore, speed up the homogenizing process. The above consideration helps to explain why it is necessary to use longer soaking times or both longer times and higher temperatures in normalizing or annealing steel castings than in the case of wrought steel of the same analysis. Homogenization is especially important in castings of heavy sections. While light castings and small test specimens such as 1.25 to 1.5 in. sq. bars may have a comparatively fine grain as cast and respond readily to grain-refining heat treatments, it is quite another problem with heavy castings which may require first a

* These structures and their causes have been discussed in considerable detail in Volume I, Chapters VIII and XI, of this monograph.

homogenizing treatment to overcome segregation and a further treatment at a lower temperature for grain refinement.

19. Distribution of Inclusions.*—Not only is the above type of segregation important in determining the properties of steel castings, but the rejection of solid non-metallic inclusions to the grain boundaries or the spaces between the dendrites may also be highly important. If such inclusions are uniformly distributed and in rounded masses, they may be comparatively harmless, but if they occur as films on the grains or as more-or-less continuous stringers around the grains, they may materially reduce the ductility and in some cases make it impossible to overcome their harmful effects by any heat treatment. The form and distribution of these defects and of gas cavities seem to be particularly important in determining the ductility of steel castings as cast and their response to heat treatment. These relations were studied by Sims and Lillieqvist,[571] Harper and Stein,[122] McCrae and Dowdell,[391,392] Lorenz,[268,543] and others.

20. Shrinkage Cavities.—The solidification of steel results in a decrease in volume—about 5.5 per cent according to experiments by Benedicks, Berlin, and Phragmén[114] and calculations by Honda and Endo[213]—and since the casting solidifies from the outside inward it is evident that molten metal must be fed to the casting to compensate for this shrinkage, or there will be porosity at the center of the casting. The foundryman recognizes these conditions and intends to provide an adequate supply of molten metal by using suitable heads of metal in risers attached to the sections which will be last to solidify. Briggs and Gezelius[594] attributed "internal hot tears" to solidification shrinkage. Shrinkage, cavities, and porosity cannot be corrected by heat treatment. It is important that castings be sound after solidification is complete, and that they do not contain excessive amounts of inclusions distributed so as to affect seriously the ductility of the casting or of such character and distribution that their embrittling effects cannot be removed by suitable heat treatments.

21. Gas Evolution.—Gas evolution at some stage in the solidification of steel castings may also be a source of trouble. While steel for castings is intended to be thoroughly killed before pouring, there is evidence to support the belief that certain gases are more soluble in the molten than in the solid metal and that they are

* See Volume I, Chapter XI, for a detailed discussion of inclusions.

liberated during solidification. Hydrogen and carbon monoxide or gases formed by their reaction with oxides in the metal are the most likely offenders. Entrapped air may also produce porosity. The evolution of gas, even in small amounts, during the solidification of a casting must result in entrapping some of the gas, thus producing voids in the final castings. Sims[772] has recently discussed the causes of the porosity which are ascribable to the metal itself. Gas evolved from the action of the hot metal on the mold causes surface defects such as pin holes. Pin-hole trouble is more common in green-sand than in dry-sand molds.

22. Recrystallization and Grain Refinement.—The recrystallization of steel castings takes place as they are cooled through the critical range, as has been outlined in its general aspects in Volume I.* At the upper critical temperature (Ar_3) the separation of the ferrite begins, and if the cooling is sufficiently slow, the ferrite forms at the boundaries of the original austenite grains, forming a continuous network, while pearlite forms within the austenite grains when the lower critical temperature is reached. This relation applies to the cooling of the casting in the mold or to any subsequent cooling from above the upper critical temperature. The size of the network structure depends upon the size of the original austenite grains from which it forms. This type of structure is characteristic of castings which have been annealed and cooled in the furnace.

If the cooling rate through the critical range is sufficiently rapid, the ferrite separates along crystallographic planes in the austenite grain and gives rise to what is known as the Widmanstätten structure.† Material showing this structure may have higher tensile and yield strengths but is likely to be lower in ductility. It is generally found in the original casting and in castings which have been normalized, although with light castings and especially with those of higher carbon percentages normalizing may prevent extensive separation of the proeutectoid ferrite and may produce a fine pearlitic structure. When high ductility is required, it is customary to reheat these castings to just below or even into the critical range, which spheroidizes the cementite

* See Chapter II, section 27; Chapter IX, section 111; Chapter XII, section 152.

† Described in detail in Volume I, Chapter VIII, section 107, of this monograph.

and greatly increases both the static and impact ductilities. The advantages of such reheating treatments have been emphasized by MacPherran and Harper,[129] Harper and Stein,[122] Merten,[273] Mitchell,[393] Lorenz,[268] and others.

Repeated heating and cooling of castings through the critical range tends to further refinement of the grain, and as is pointed out in section 11 of this chapter, double and triple heat treatments are employed in order to meet special requirements.

When the maximum grain refinement and optimum mechanical properties of steel castings are sought, a water or oil quench from above the upper critical temperature should be employed when the design and character of the casting permit such a treatment without danger of cracking or undue stressing. The use of a liquid quench for castings of medium carbon is especially beneficial, because these castings show marked improvement in properties and are less likely to be damaged than the higher carbon castings. Some investigators maintain that all castings which are to be subjected to a liquid quench should first be thoroughly annealed; other authorities are not in agreement with this. The use of hot water and the removal of the castings from the quenching bath while still hot are sometimes recommended when there is danger of cracking. Most castings subjected to a liquid quench should be tempered, the tempering temperature depending upon the requirements with reference to strength and ductility. For maximum ductility the tempering should be just below the critical range.

Hatfield[709] showed, for two cast carbon steels, the variation in properties, especially in ductility, which may be produced by variations in heat treatment. His data are given on pages 55, 56, 63, and 64.

23. Shrinkage of Steel Castings.—There are three kinds of shrinkage involved in the production of steel castings: (1) decrease in the volume of the molten metal before solidification begins, (2) decrease in volume during solidification, and (3) the net contraction of the solidified casting as it cools to room temperature.

The first and second should be provided for in good foundry practice by keeping the mold full of molten metal and by feeding molten metal to the last sections to solidify until solidification is completed. The contraction of the casting during cooling is

unavoidable and must be provided for in the pattern. Since steel solidifies at a much higher temperature than cast iron—1530 to 1480°C. (2785 to 2695°F.) as compared with 1300 to 1150°C. (2370 to 2100°F.)—and since there is no graphite formation, it follows that steel castings will show greater shrinkage. It is common foundry practice to provide for a pattern shrinkage of ¼ in. per ft. for steel castings as contrasted with ⅛ to 5/32 in. per ft. for cast irons.

A comprehensive study of the shrinkage of steel castings has been reported by Körber and Schitzkowski,[263] who determined the shrinkage of specimens 450 mm. (17.7 in.) in length and 20, 30, 40, and 50 mm. (0.79, 1.18, 1.57, and 1.97 in.) in diameter. The method used in determining the shrinkage was that previously described by Wüst and Schitzkowski[96] and provided for automatic recording of the shrinkage curve; the temperature of the specimens was determined by placing a platinum–platinum-rhodium thermocouple in the gate just above the specimen.

The compositions of the steels investigated were reported as follows:

Steel	Composition ranges, per cent				
	C	Mn	Si	P	S
Acid open-hearth	0.28 to 0.50	0.26 to 0.69	0.22 to 0.60	0.063 to 0.106	0.036 to 0.085
Basic open-hearth	0.15 to 0.39	0.49 to 0.80	0.32 to 0.47	0.014 to 0.047	0.021 to 0.045

Three series of tests were made on the acid steels and two on the basic, as follows:

Series 1. Shrinkage determinations on straight round bars of acid steel, cast in green-sand molds.

Series 2. Tests on bars with end plates or flanges about 100 mm. (3.94 in.) greater in diameter than the bars. The *A*, *B*, *C*, and *D* specimens were cast in green-sand molds, which offered little resistance to contraction, while the *A′*, *B′*, *C′*, and *D′* specimens were cast in a special dry mold, which offered high resistance to contraction of the specimens. Both acid and basic steels were used.

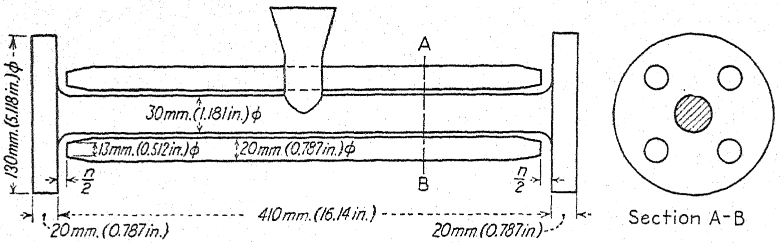

Fig. 1.—Method of limiting shrinkage in cast-steel bars. (*Körber and Schitzkowski.*[263])

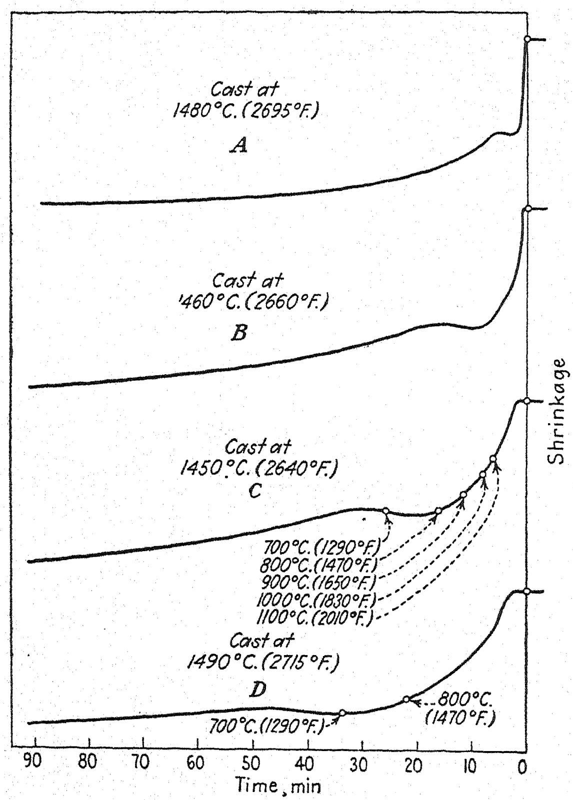

Fig. 2.—Shrinkage-versus-time curves for cast-steel bars 450 mm. (17.7 in.) long (series 1) with the following diameters: *A*—20 mm. (0.79 in.), *B*—30 mm. (1.18 in.), *C*—40 mm. (1.57 in.), and *D*—50 mm. (1.97 in.). (*Körber and Schitzkowski.*[263])

Series 3. A round bar with end flanges was cast with the dimensions shown in Fig. 1. Four round rods were placed between the flanges with a clearance which would permit a definite contraction. This contraction, n, was 2, 4, 6, 8, 10, and 12 mm. (0.078, 0.156, 0.234, 0.315, 0.394, and 0.472 in.). The rods were placed so that one-half of the contraction, $n/2$, occurred at each flange. Both acid and basic steels were used.

The results of the tests of series 1 are given in Table 2 and shown in Fig. 2. It is evident that the smaller bars contract much more rapidly, but all show the same final contraction. The more rapid contraction of the smaller sections has an important bearing on hot-shrinkage cracks in castings of irregular section. The contraction above and below the lower critical temperature was about equal.

TABLE 2.—SHRINKAGE OF ROUND-BAR CASTINGS OF ACID OPEN-HEARTH STEEL. SERIES 1*

Diameter		Shrinkage, per cent			Casting temperature	
mm.	in.	Total	To Ar_1	Below Ar_1	°C.	°F.
20	0.79	2.20	1.19	1.01	1480	2695
30	1.18	2.18	1.21	0.97	1460	2660
40	1.57	2.13	1.06	1.07	1450	2640
50	1.97	2.16	1.20	0.96	1490	2715

* Körber and Schitzkowski.(263)

The results of the tests of series 2 are given in Table 3. It is significant that, when the cast bar is not free to contract during cooling, cracks form, and in all these cases fractures result. Even when about 80 per cent of the normal contraction took place, one specimen fractured. These data also have an important bearing on the danger of cracking small or light sections between larger ones. The danger of cracking the casting when the mold offers high resistance to shrinkage of the casting is also evident.

The results of tests of series 3 are not reported in detail here, but they showed that, unless considerable contraction was provided for, the castings fractured; the minimum percentage of shrinkage which did not cause fractures was 1.75 for the acid steel and 1.54 for the basic steel.

Over the range of casting temperatures of 1450 to 1510°C. (2640 to 2750°F.), there was only slight variation in the shrinkage of the casting, but the tendency was for the shrinkage to increase as the casting temperatures increased. No difference between the shrinkage of the acid and the basic steels was noted. A shrinkage

TABLE 3.—SHRINKAGE OF CAST BARS WITH FLANGED ENDS. SERIES 2*

Diameter, mm.	Diameter, in.	Specimen†	Shrinkage, per cent	Casting temperature, °C.	Casting temperature, °F.	Condition of casting
			Acid open-hearth steel			
20	0.79	*A*	2.03	1470	2680	Not broken
		A′	0.34	1460	2660	Broken
30	1.18	*B*	2.13	1500	2730	Not broken
		B′	0.61	1470	2680	Broken
40	1.57	*C*	2.07	1480	2695	Not broken
		C′	0.89	1460	2660	Broken
50	1.97	*D*	2.19	1470	2680	Not broken
		D′	1.78	1490	2715	Broken
			Basic open-hearth steel			
20	0.79	*A*	1.99	1480	2695	Not broken
		A′	0.55	1450	2640	Broken
30	1.18	*B*	2.10	1460	2660	Not broken
		B′	0.43	1440	2625	Broken
40	1.57	*C*	2.10	1500	2730	Not broken
		C′	0.50	1500	2730	Broken
50	1.97	*D*	2.08	1480	2695	Not broken
		D′	0.91	1485	2705	Broken
100	3.94	*E*	2.20	1510	2750	Not broken

* Körber and Schitzkowski.(263)

† Specimens *A*, *B*, *C*, *D*, cast in green-sand molds. Specimens *A′*, *B′*, *C′*, *D′*, cast in special dry molds.

of ¼ in. per ft. corresponds to 2.08 per cent, which is in good agreement with the average values reported in these tests.

The effect of chemical composition on the shrinkage of steel castings has been summarized by Körber and Schitzkowski[263] as follows:

0.1 per cent carbon decreases shrinkage 0.03 per cent.
0.1 per cent manganese increases shrinkage 0.006 per cent.
0.1 per cent silicon increases shrinkage 0.011 per cent.
0.1 per cent phosphorus decreases shrinkage 0.031 per cent.
0.1 per cent sulphur decreases shrinkage 0.034 per cent.

The above data were considered to apply up to the maximum values of carbon 0.5, manganese 1.0, silicon 0.8, phosphorus 0.1, and sulphur 0.1 per cent, and it is obvious that for the variations in the compositions found in steel castings the chemical composition has only a minor effect on the shrinkage. While the chemical composition may have some influence on the propensity of the castings to show shrinkage cracks, it appears that the design and the molding practice may be of greater importance.

Briggs and Gezelius[594] studied the solid shrinkage of steel castings and concluded that "external hot tears" are due to the stresses introduced during the early stages of the solid shrinkage and that such cracks occur at temperatures above about 705°C. (1300°F.).

The First Report of the Steel Castings Research Committee[623] devoted much space to the importance of shrinkage in various forms, particularly with reference to the effect of design. It stated: "In the present state of development of the steel castings industry it is therefore not too much to say that the next essential step in progress is in the design of castings."

The importance of design and its relation to shrinkage and the production of sound castings were emphasized by Briggs and Gezelius,[594] Heuvers,[305] Lorenz,[544] and Bull.[444]

D. EFFECT OF CHEMICAL COMPOSITION ON PROPERTIES

The chemical composition of the steel determines the potential properties of the casting. There are so many other contributing factors, however, that it is not possible to predict the mechanical properties of castings in either the cast or the heat-treated condition. The factors other than chemical composition which influence the mechanical properties of steel castings may be grouped

under the general heading of foundry practice, which includes character of the material charged into the furnace, melting and refining practice—deoxidation, recarburization, pouring temperature, special addition agents—kind of molds, gating, risers, and

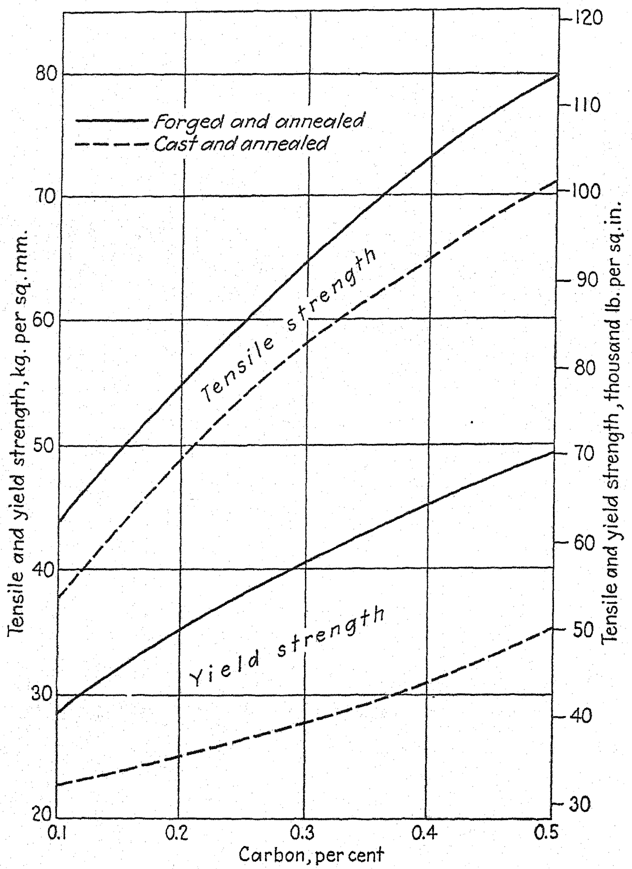

FIG. 3.—Effect of carbon content on tensile and yield strengths of annealed cast and wrought steels. (*Delbart.*[601])

finally the size and design of the casting or test coupon. There are some differences due to the melting process, but frequently other factors are more important.

24. Carbon.—If other conditions are equivalent, the tensile and yield strengths, proportional limit, and hardness of cast steels increase as the carbon content increases. This increase is almost

directly proportional to the carbon content, but the relation is not exactly linear so that it cannot be represented accurately by a simple equation. The effectiveness of heat treatment in increasing strength also depends upon the carbon content; for example, the increased strength resulting from hardening and tempering a 0.50 per cent carbon steel is much greater than the increase in strength of a 0.25 per cent carbon steel similarly treated.

Delbart(601) compared the tensile and yield strengths of wrought and cast steels as a function of the carbon content (up to 0.50 per cent carbon); his curves are reproduced in Fig. 3. The castings were cast in 40-mm. (1.57-in.) blocks in dry-sand molds. The other elements in the cast steels were of nominal amounts (0.25 to 0.35 per cent silicon, 0.60 to 0.80 per cent manganese, and 0.050 per cent phosphorus and sulphur). Both wrought and cast specimens were slowly cooled in the furnace from the annealing temperature. Further information regarding the effect of the carbon content on the mechanical properties of steel castings may be had by comparing the properties reported in Tables 9, 11, 14, 16, 20, and 23 of Chapter III.

Unsoundness is generally more prevalent in the low-carbon castings (not over 0.21 per cent carbon; see page 52) if together with the low carbon there are also low manganese and silicon. Increasing the carbon content in steel castings generally results also in decreasing the ductility as indicated by lower elongation, reduction of area, and impact resistance. These effects are more pronounced in the higher carbon (above 0.35 per cent) castings in the cast or normalized condition. Proper heat treatment, especially reheating slightly below the lower critical point, aids materially in improving the ductility of these higher carbon steel castings.

25. Manganese.—Manganese is recognized as an element which contributes to soundness of castings, and at least a nominal manganese content (0.30 to 0.50 per cent) is generally considered necessary, although it has been shown by Becket(435) that sound steel can be produced without manganese by the use of some substitute such as zirconium. As stated on page 25, the manganese content of castings is frequently not specified, but in some cases a maximum of 0.80 to 0.85 per cent is indicated, while in one instance (A.S.T.M. Specification A95-29) a minimum

of 0.50 per cent is given. Bull[512] expressed a preference for minimum rather than maximum values for manganese, contending that at least certain amounts are desirable for safeguarding the soundness of the castings. In larger amounts, manganese may be considered as an alloying element, as in the case of steels containing from 1 to 1.5 per cent manganese. Even when the manganese is increased only to the higher amounts found in carbon-steel castings it increases the strength. It is good practice to consider the sum of the manganese and carbon and to avoid running high on both in a given heat of steel. Manganese is further beneficial in counteracting certain harmful effects of sulphur.

26. Silicon.—Silicon, like manganese, contributes to the soundness of castings, being considered a more active deoxidizing agent. Specifications usually permit the foundryman to vary the silicon content, but maximum (0.40 per cent) or minimum (0.20 per cent) contents are specified in a few instances. Silicon, when enough is present, may increase the strength of the casting. Merten,[273] in a paper on large castings, reported that the best properties were obtained when the final silicon was below 0.20 per cent, but Bull, in discussing this report, maintained that good castings were obtained with higher silicon contents and quoted White,[171] who had formerly recommended that the silicon in steel castings should never be less than 0.20 per cent. Thus the function of silicon in steel castings seems to be understood, but the optimum amount seems to be debatable, probably depending upon other conditions, such as the amounts of carbon and manganese present and possibly upon the size of the casting.

27. Phosphorus.—The amount of phosphorus permitted in steel castings is somewhat higher than in the corresponding wrought material and ranges from 0.050 to 0.080 per cent. Low phosphorus is generally recommended as a means of safeguarding against cold-shortness in steel castings. Langenberg,[73,74] as a result of tests on normalized and tempered and on quenched and tempered specimens, concluded that increasing the phosphorus affected the impact resistance more than tensile properties and that there was a pronounced drop in the impact resistance between 0.060 and 0.070 per cent phosphorus. Bull[512] favors holding the phosphorus to 0.050 per cent maximum, until it is

established that higher amounts can be used with safety.* In basic-furnace practice there should be no difficulty in keeping the phosphorus below this limit. Fenstermacher(449) reported on the analyses of 59 heats of open-hearth carbon-steel castings produced during a single month, in which the phosphorus ranged from 0.048 to 0.012 per cent with an average of 0.022 per cent. In acid-furnace practice more care must be exercised in the selection of the melting stock in order to keep the phosphorus within specification limits, which are sometimes slightly higher than for basic steel.

Phosphorus tends to raise tensile and yield strengths with corresponding reduction in elongation and reduction of area.

28. Sulphur.—The specification maxima for sulphur range from 0.050 to 0.080 per cent. The object in keeping down the sulphur content of steel castings is to safeguard against red-shortness and segregation. Since the Joint Committee on Investigation of the Effect of Phosphorus and Sulphur in Steel(79) reported that it could find no relationship between serviceability and a sulphur content in the steel as high as 0.09 per cent, some question may be raised regarding the necessity for some of the present requirements, and Bull expressed the opinion that soon the maximum will be set at 0.06 per cent in such accredited specifications as now set the maximum limit at less than 0.06 per cent.

Both phosphorus and sulphur must be broadly regarded as objectionable impurities in steels, but their limits should be established on an economic basis of the cost of materials and the required properties.

29. Other Elements.—Other elements which may be present in carbon-steel castings within the limits set for this monograph may be considered to have negligible effects on the mechanical properties. A possible exception must be made in the case of

* Gillett,(798) in a comprehensive summary on phosphorus as an alloying element in steel, concluded that (in wrought steels) "the exact mechanism of phosphorus embrittlement remains obscure, and one cannot yet predict the phosphorus limit in an alloy steel within which it is satisfactorily tough and beyond which it is impact-brittle without actual trial . . . At some high-phosphorus content, varying according to the carbon and alloy composition . . . room-temperature impact resistance will drop, and this is still more marked at subnormal temperatures. This occurs at phosphorus contents below those at which static ductility or formability are seriously affected."

aluminum, not because of the presence of metallic aluminum but because of the effects of its reaction products. The effect of aluminum additions on the mechanical properties of steel castings has been studied by many metallurgists and foundrymen.

Lorenz, in discussion of McCrae and Dowdell's paper,[392] reported, as a result of 60 tests on specimens of open-hearth steel from five heats containing from 0.25 to 0.35 per cent carbon and from 0.77 to 1.50 per cent manganese, that the addition of 1 oz. of aluminum per ton lowered the ductility but had little effect on the strength. All samples were heat treated at 900°C. (1650°F.), one group being furnace cooled, another air cooled, and the third water quenched and then tempered at 650°C. (1200°F.). The elongation, reduction of area, and Charpy values were 22.95, 31.4, and 13.30 respectively for those with the aluminum addition and 27.45, 47.3, and 20.45 for those without the aluminum.

On the other hand, McCrae and Dowdell[391,392] after numerous experiments with acid electric steel concluded that aluminum was equal or superior to the other deoxidizers tested in producing sound castings and that if the carbon, manganese, and silicon were held within certain limits (0.10 to 0.17 per cent carbon, 0.60 to 0.70 per cent manganese, and 0.30 to 0.40 per cent silicon), about 1 lb. of aluminum per ton of steel could be added to the ladle shortly before pouring without lowering the ductility of normalized castings below the usual specified limits. Hamilton, in discussing a paper by McCrae and Dowdell,[392] described an acid electric-furnace practice by which sound castings were obtained with the addition of 10 oz. of aluminum per ton of steel. He reported typical analysis and properties after normalizing at 900°C. (1650°F.) as follows: Analysis—0.23 per cent carbon, 0.69 per cent manganese, 0.40 per cent silicon, 0.029 per cent phosphorus, 0.034 per cent sulphur. Properties—tensile strength 74,800 lb. per sq. in., yield strength 43,200 lb. per sq. in., elongation, in 2 in., 29.0 per cent, and reduction of area 50 per cent.

Sims and Lillieqvist[571] emphasized the importance of maintaining the proper relation between the oxides to be reduced and the amount of aluminum or other strong deoxidizer added, in order to avoid the formation of oxidation products or sulphides in a eutectic pattern.

Apparently still further work is needed on the subject of deoxidation, sound castings, and low ductility in castings.

E. EFFECT OF SIZE OF CASTING AND LOCATION OF TEST SPECIMEN ON MECHANICAL PROPERTIES

It is generally conceded that better mechanical properties can be obtained in castings of small section than in large ones and that in the latter the core of the casting will be inferior in properties to the material nearer the surface. Moreover, as has been stated on a previous page, it is more difficult to heat treat larger castings so that optimum properties will be obtained. These factors have been studied by Heinrich, Merten, Lorenz. and by Harper and associates.

30. Heinrich's Investigation.—Heinrich[530] used a converter steel of the following composition: 0.08 to 0.11 per cent carbon, 0.45 to 0.60 per cent manganese, 0.20 to 0.30 per cent silicon, 0.05 to 0.07 per cent sulphur, and 0.06 to 0.08 per cent phosphorus. He had previously found that specimens cut from axle housings frequently had strength properties which were lower than those specified. Because of the shape of the housing the specimen had to be cut so that the parting line of the mold practically bisected the specimen. Heinrich was, therefore, interested in two variables: first, the properties of specimens cut from the casting as compared with specimens cast separately, and second, the effect of slightly shifting the parting line on the mechanical properties. The latter part of the investigation, for which a basic electric steel was used, is not of interest here except for the fact that the location and the shifting of the parting line do deleteriously affect the properties.

In investigating the properties of specimens cut from the casting and of specimens separately cast, Heinrich first summarized the tensile strength and elongation of 1108 specimens cut from axle housings as cast in regular production over a two-year period. Statistical curves for these are shown at the top of Fig. 4. He next summarized the properties of 100 specimens cut from selected axle housings which were apparently satisfactory, and of 100 specimens from separately cast blocks. The statistical curves for these are shown at the center and bottom of Fig. 4.

The axle housings and the separately cast blocks were left in the sand 5 to 7 hr. and were then annealed at 930°C. (1705°F.).

The annealing time was 1.5 hr. for the housings (which had a relatively thin wall) and 2 hr. for the blocks. The housings and the blocks were cooled in air to 700°C. (1290°F.), returned to the annealing furnace, and cooled in the furnace to 100°C. (210°F.) in 6 hr. Specimens were 10 mm. (0.39 in.) in diameter and of 100 mm. (3.94 in.) test length.

Heinrich concluded that (as is evident from Fig. 4) casting variables affect the structure and properties of specimens cut

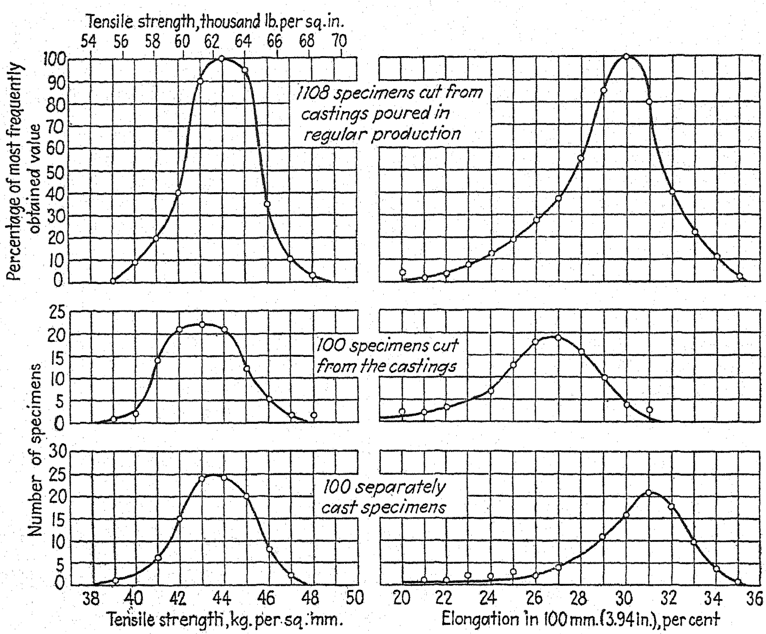

FIG. 4.—Tensile strength and elongation of specimens cut from annealed axle-housing castings and of annealed separately cast specimens. (*Heinrich.*[530])

from castings so much that it is only by testing specimens cast separately that properties can be properly evaluated.

31. Investigations of Merten and of Lorenz.—Merten's[273] results show the effect of size of the coupon from which the test specimen was taken and the effect of the following heat treatments on the properties of large castings:

Heat Treatment A. Normalized at 980°C. (1800°F.) for 36 hr. and cooled to 650°C. (1200°F.). Reheated at 850°C. (1565°F.) for 20 hr. and cooled in the furnace to 250°C. (400°F.).

Heat Treatment B. Normalized at 1100°C. (2010°F.) for 6 hr., reheated at 860°C. (1580°F.) for 6 hr., and cooled in air to 315°C. (600°F.).

The steel contained 0.25 per cent carbon, 0.49 per cent manganese, 0.30 per cent silicon, 0.034 per cent sulphur, and 0.04 per cent phosphorus. The properties are shown in Table 4.

TABLE 4.—EFFECT OF SIZE OF TEST COUPON AND OF HEAT TREATMENT ON MECHANICAL PROPERTIES OF STEEL CASTINGS*

Test coupon	Heat treatment (see text)	Tensile strength, lb. per sq. in.	Yield strength, lb. per sq. in.	Elongation in 2 in., per cent	Reduction of area, per cent	Remarks
1.5 × 2 × 6 in.	*A*	70,800	44,900	30	49	Average of 4 tests
12 × 14 × 19 in.	*A*	61,000	30,000	16	26	Single test
12 × 14 × 19 in.†	*B*	66,150	33,800	26.7	42	Average of 6 tests
12 × 14 × 19 in.‡	*B*	66,000	33,050	25.8	38	Average of 5 tests
12 × 14 × 19 in.§	*B*	59,500	27,150	34.1	55.7	Average of 6 tests

* Merten.[273]

† Tests on core-drilled specimens from surface to 5-in. depth.

‡ Tests on core-drilled specimens from center of coupon.

§ Tests on core-drilled specimens from surface to 5-in. depth but with different prolongation.

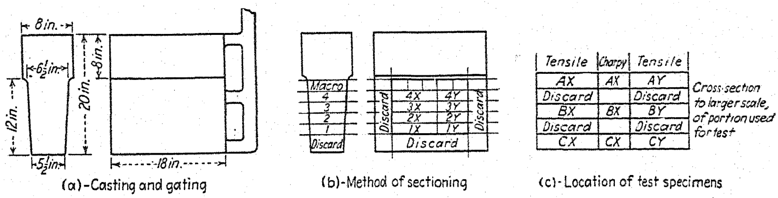

FIG. 5.—Method of casting and location of specimens in block. (*Lorenz.*[543])

It will be noted from this table that, by using a high normalizing temperature followed by a refining treatment at a lower temperature, Merten was able to produce good mechanical properties in specimens cut from large test coupons, even when the specimen was cut from the center section. From these data and others not quoted here, Merten concluded that core-drilled specimens from critical sections of heavy castings or prolongations of comparable sections should be tested, and that with

proper heat treatment such specimens can be made to pass specifications.

Lorenz[543] determined the mechanical properties of specimens from different sections of the 6-in. block shown in Fig. 5, using four different heat treatments. The locations of the specimens are also shown in Fig. 5. It will be noted that the *A* and *C* specimens were from the outside of the casting, while tne *B* specimens were from the central section of the block. The various slabs were heat treated in sections approximately 1 in. square and 14 in. in length. The steel contained 0.28 per cent carbon, 0.87 per cent manganese, 0.36 per cent silicon, 0.046 per cent sulphur, and 0.021 per cent phosphorus. The heat treatments are given in Table 5 and the resulting mechanical properties in

TABLE 5.—HEAT TREATMENT OF SLABS FROM CAST 6-IN. BLOCK*

Specimen	No.	Heat treatment
1*X* and 4*Y*	1	Heated to 900°C. (1650°F.), held 2.5 hr., air cooled. Tempered at 595°C. (1100°F.) for 5 hr.
4*X* and 1*Y*	2	Heated to 900°C. (1650°F.), held 12 hr., air cooled. Tempered at 595°C. (1100°F.) for 5 hr.
2*X* and 3*Y*	3	Heated to 995°C. (1825°F.), held 12 hr., air cooled. Heated to 870°C. (1600°F.), held 2 hr., air cooled. Tempered at 595°C. (1100°F.) for 5 hr.
3*X* and 2*Y*	4	Heated to 995°C. (1825°F.), held 12 hr., air cooled. Tempered at 595°C. (1100°F.) for 5 hr.

* Lorenz.[543]

Table 6. The significant results of these tests are the good properties of the specimens from the central section of the casting, the tensile strength being about 95 per cent of that of specimens from the outside sections, while the elongation, reduction of area, and Charpy impact values compared still more favorably. No significant differences in the properties resulted from these rather widely differing heat treatments. Thus, increasing the time at 900°C. (1650°F.) from 2.5 to 12 hr. or raising the temperature from 900 to 995°C. (1650 to 1825°F.) had no important effect on the resulting properties.

32. Investigations of Harper and Associates.—In the work by Harper and Stein[122] particular attention was paid to the properties of test-coupon bars as compared with drilled specimens

TABLE 6.—MECHANICAL PROPERTIES OF SPECIMENS FROM DIFFERENT SECTIONS OF 6-IN. CASTING*

Specimen	Treatment	Tensile strength, lb. per sq. in.	Yield strength, lb. per sq. in.	Elongation in 2 in., per cent	Reduction of area, per cent	Charpy impact, ft-lb.
1-*AX*	1	75,750	45,650	31.5	46.9	18.92
4-*AY*	1	74,300	45,200	27.5	42.8	
4-*AX*	2	75,450	47,950	23.5	36.6	19.18
1-*AY*	2	75,000	48,950	26.5	38.8	
2-*AX*	3	74,300	46,200	28.5	44.0	20.38
3-*AY*	3	74,600	50,050	29.5	43.4	
3-*AX*	4	74,400	50,450	30.0	48.9	19.45
2-*AY*	4	74,100	47,400	31.5	41.9	
1-*BX*	1	71,350	45,075	28.5	48.3	24.71
4-*BY*	1	71,900	45,900	23.5	38.2	
4-*BX*	2	71,900	43,450	27.0	36.3	27.76
1-*BY*	2	71,700	43,800	29.5	46.3	
2-*BX*	3	71,100	46,250	29.0	42.8	24.56
3-*BY*	3	71,550	43,800	31.5	48.6	
3-*BX*	4	71,350	46,750	30.0	41.9	30.77
2-*BY*	4	71,300	46,300	31.0	45.4	
1-*CX*	1	75,550	48,350	27.5	43.1	20.24
4-*CY*	1	75,150	49,100	25.0	43.4	
4-*CX*†	2	76,650	47,350	24.0	22.7†	20.24
1-*CY*	2	74,950	46,250	23.5	31.5	
2-*CX*	3	74,700	49,050	30.0	44.9	24.85
3-*CY*	3	74,400	45,400	25.0	32.1	
3-*CX*	4	74,100	46,850	26.2	36.9	22.02
2-*CY*	4	72,900	49,350	29.5	46.0	

* Lorenz.[543]

† Test specimen machined incorrectly to 0.485 in.

from castings weighing from 1000 to 20,000 lb. and to the effect of single, double, and triple heat treatments. Details of the specimens and the heat treatments are as follows:

Steel	Specimen and Heat Treatment
A.........	Test-coupon bar, 1 × 1 × 6 in., on 1000-lb. casting, annealed 8 hr. at 870°C. (1600°F.).
B.........	Specimen drilled from 1000-lb. casting after heating 2 hr. at 845°C. (1550°F.), quenching in

water to 650°C. (1200°F.), followed by annealing 4 hr. at 830°C. (1525°F.).

C......... Specimen drilled from 20,000-lb. casting. Normalized 15 hr. at 950°C. (1750°F.), annealed 24 hr. at 845°C. (1550°F.).

D......... Specimen drilled from 12,000-lb. casting after normalizing 10 hr. at 885°C. (1625°F.), annealing 10 hr. at 845°C. (1550°F.), reheating 3 hr. at 695°C. (1290°F.).

The data for these specimens are given in Table 7.

Of special interest is the higher ductility as shown by the elongation and reduction of area of the specimens which were quenched or normalized as compared with the specimen which was simply annealed. It will be noted that the specimens drilled from the large casting which had been given proper heat treatments showed physical properties which compared favorably with the specimens from test-coupon bars.

MacPherran and Harper[129] cited data on test bars core drilled from 10-ton castings containing about 0.30 per cent carbon, 0.70 per cent manganese, 0.30 per cent silicon, 0.035 per cent sulphur, and 0.04 per cent phosphorus, held 10 hr. at 900°C. (1650°F.), air cooled, held 10 hr. at 845°C. (1550°F.), furnace cooled, then given a 3-hr. spheroidizing treatment at 700°C. (1290°F.), and air cooled. These castings showed, on specimens taken radially, 70,000 to 75,000 lb. per sq. in. tensile strength and 36,500 to 42,000 lb. per sq. in. yield strength, with 30 per cent elongation (gage length not given, probably 2 in.) and 46 to 50 per cent reduction of area.

TABLE 7.—COMPARISON OF PROPERTIES OF COUPONS AND DRILLED SPECIMENS OF REGULAR-GRADE STEEL CASTINGS*

Steel	Composition, per cent					Tensile strength, lb. per sq. in.	Yield strength, lb. per sq. in.	Elongation in 2 in., per cent	Reduction of area, per cent
	C	Mn	Si	P	S				
A	0.28	0.60	0.29	0.035	0.044	71,300	37,700	20	28
B	(Same as *A*)					66,850	37,900	30	45
C	0.28	0.67	0.31	0.034	0.040	74,900	43,700	30	50
D	0.31	0.71	0.29	0.034	0.039	70,000	36,500	30	50

* Harper and Stein.[122]

They also reported on core-drilled radial specimens from a 50-ton casting of 0.28 per cent carbon, 0.69 per cent manganese, 0.33 per cent silicon, 0.037 per cent sulphur, and 0.032 per cent phosphorus steel, held 20 hr. at 870°C. (1600°F.) and furnace cooled, then held at 675°C. (1250°F.) for 18 hr., for spheroidizing, and furnace cooled, which gave 66,000 lb. per sq. in. and 30,500 lb. per sq. in. tensile and yield strengths with 32 per cent elongation and 49 per cent reduction of area.

F. AUTHOR'S SUMMARY

1. The most important recent developments in the steel-castings industry have been the introduction of modern methods of heat treatment and the use of alloy steels. Most steel castings are now given some form of heat treatment. There are many modifications of these heat treatments, such as annealing, normalizing, normalizing followed by annealing, double annealing, quenching in a liquid medium, annealing or normalizing followed by quenching in a liquid medium, etc. It is general practice to follow normalizing or quenching in a liquid medium by tempering, the temperature used in tempering depending upon the properties required and the composition of the steel, but for maximum toughness it is customary to heat to just below the lower critical point. There is a marked increase in the use of a liquid quench as a means of developing the highest possible mechanical properties, and this method of heat treating makes possible a wide range in the strength and toughness properties by varying the tempering temperature and time.

2. Specifications for steel castings of different grades or classes have been developed by technical societies and by the larger consuming industries. These specifications have aided in standardizing the quality of castings of different classes, but there are indications that some of them need revision in order to make them more nearly representative of modern industrial practice.

3. There is a lack of agreement or standardization in regard to casting separate test specimens, the use of coupons attached to the castings, or of specimens cut or drilled from the castings. The use of lugs or coupons attached to the castings is more extensively employed in present practice as a means of providing test specimens.

4. The temperatures used in normalizing and annealing vary considerably in different plants, ranging from about 870 to 1090°C. (1600 to 2000°F.). In general, higher temperatures are used for the larger castings and in those heat treatments which are intended to remove segregation or cast structure. Lower temperatures are used to effect the maximum grain refinement. A combination of two heat treatments may be used, in which the first is at a higher temperature for homogenizing the casting, and the latter at a lower temperature for maximum grain refinement. Normalizing or annealing prior to heating for liquid quenching is favored by some metallurgists but is not considered necessary by others and is not always used in practice.

5. The time of holding the castings at the heat-treating temperature (soaking) is generally not less than 1 hr. for each inch in maximum thickness of section, and times up to 2 hr. or more per inch of thickness of maximum section are used.

6. Radiography plays a rather important part in detecting defects in steel castings, in helping to eliminate certain defects, and in providing a non-destructive method of testing materials which are to be used for special service.

7. The solidification shrinkage of steel castings amounts to 5 to 6 per cent of the volume, and when the castings are not properly gated and fed may cause shrinkage cavities or internal hot tears.

8. In cooling from the solidification temperature to ordinary temperatures, cast steel shows a linear shrinkage of slightly over 2 per cent, which is known as pattern shrinkage, and foundrymen usually provide for a linear shrinkage of ¼ in. per ft., which corresponds to 2.08 per cent. Solid shrinkage is thought to cause external hot tears when the mold offers too much resistance to shrinkage of the casting or when the design introduces severe stresses due to unequal rates of cooling.

9. The chemical composition determines the potential properties of castings, but foundry practice and heat treatment may be more important factors. As a result, chemical composition is used less as a basis for specifications or purchases than are mechanical properties. The functions of the various elements and their usual ranges in percentage have been given.

10. The form and distribution of inclusions in steel castings are known to be of great importance in determining their ductility.

There is general agreement that inclusions in rounded form randomly distributed are less harmful than when they occur in stringers or in eutectic-like pattern at the grain boundaries. The exact conditions with reference to chemical composition and foundry practice which cause some melts of steel to produce castings of low ductility while others of apparently the same analysis produce castings of good ductility are not definitely known; this subject seems to require further study.

11. Data have been presented to show the effect of size of casting and location of the test specimen on the mechanical properties. Lower values are found in the larger castings, but with proper foundry practice and subsequent heat treatment the properties of specimens from heavy sections can, in some cases, be made to approach rather closely those of coupons attached to the castings or even of separately cast small bars.

CHAPTER III

MECHANICAL PROPERTIES OF CAST STEELS

Low-carbon Steel Castings—Regular-grade Steel Castings—High-carbon Steel Castings—Torsional and Compressive Properties—Author's Summary

As noted in the previous chapter (page 24) the usual specifications for steel castings give only the minimum requirements for the mechanical properties of three grades—soft, medium, and hard—and the maximum allowable percentages of sulphur and phosphorus. Foundrymen generally favor the use of mechanical properties as a basis for the purchase of castings, instead of the chemical composition (*i.e.*, carbon, manganese, and silicon ranges in addition to the regularly specified maxima for sulphur and phosphorus), or a combination of properties and composition.

In discussing the properties of cast steel it is, however, advisable to group the data according to carbon content; therefore, the classification followed by Lorig and Williams(545) and advocated earlier by Bull(442) has been used. Castings containing from 0.22 to 0.35 per cent carbon have been referred to as "regular-grade carbon-steel castings" because, as is generally known, this range in carbon content includes by far the major tonnage of steel castings produced. Bull recommended that the other elements present in castings of this grade should be in the following range: 0.50 to 0.90 per cent manganese, 0.20 to 0.70 per cent silicon, 0.050 per cent maximum phosphorus, and 0.060 per cent maximum sulphur. These requirements have not been strictly adhered to in the selection of the data here presented, but in all possible cases the analysis of the steels has been given.

The two other classes are referred to as "low-carbon steel castings," in which the carbon is less than 0.22 per cent, and "high-carbon steel castings," in which the carbon ranges from 0.36 to 0.60 per cent. A very limited tonnage of steel castings has been produced in which still higher carbon contents have

been used; these may be referred to as "extra-high-carbon steel castings," although this term has not been used in industry.

In comparing very roughly the Lorig and Williams classification with the usual commercial grades, the "regular grade" of the former corresponds approximately to the medium grade. Soft commercial castings rarely contain less than 0.15 per cent carbon, and hard castings seldom more than 0.50 per cent carbon. Lorig and Williams, however, quoted properties of low-carbon castings with as little as 0.09 per cent, and high-carbon castings containing as much as 0.86 per cent carbon, some of which will be considered under the classification of extra-high-carbon castings.

The collection of data on mechanical properties of cast steel assembled by Lorig and Williams[(545)] has been liberally used in this chapter to indicate typical properties of the various grades. It is emphasized in the original paper[(545)] that published data are rarely complete in giving chemical composition, method of casting, size of casting, heat treatment, size of test specimen, and method of testing; therefore values are not always strictly comparable. Different countries use specimens of different sizes and different gage lengths; values for elongation and reduction of area, therefore, lose some of their significance. German investigators usually express impact values in terms of energy absorbed per unit of cross-sectional area, while American investigators simply report the energy absorbed for the different specimens. The only possible means of reproducing some of these data seems to be to report them just as published, because accurate conversions cannot be made.

In general, the details hereinafter quoted from Lorig and Williams'[(545)] paper, of composition and heat treatment, are fairly complete. Of the other variables mentioned above, which may affect the mechanical properties and concerning which there are no clues in the original compilation, the effect of mass is clearly the most important. It was shown in the previous chapter that, in general, lower properties are found in test bars from large castings; but with proper foundry practice and subsequent heat treatment the properties of specimens from heavy sections can in many cases be made to approach rather closely those of coupons attached to the castings or even of separately cast bars.

It may be assumed that in most cases the specimens for which properties are given in Tables 9, 11, 14, 16, and 20 were from attached coupons and are probably representative of the properties of small castings of the given composition after the specified heat treatment. It may also be assumed, especially for the higher carbon steels, that the properties of large castings would be somewhat lower than the values given in these tables.

A. MECHANICAL PROPERTIES OF LOW-CARBON STEEL CASTINGS

As noted before, Lorig and Williams classified as low carbon those cast steels which contained less than 0.22 per cent carbon. To simplify tabulation of the mechanical properties these have been subdivided further: into steels with 0.09 to 0.14 per cent, and steels with 0.15 to 0.21 per cent carbon.

33. Cast Steels Containing 0.09 to 0.14 Per Cent Carbon.—Compositions, heat treatments, and properties at normal temperature of cast steels containing 0.09 to 0.14 per cent carbon are given in Tables 8 and 9, arranged in the order of their carbon content. The values on steel 2, which indicate the effect of heat treatment, are from Hatfield;[(709)] all of the other data are from the Lorig and Williams[(545)] paper, with the original sources cited.

Low-carbon steel castings were formerly used almost exclusively in the cast condition; heat treatment was not necessary to produce fair ductility, and in many foundries no annealing facilities were available. Lately, however, heat treatment of these low-carbon grades has become more generally used; for many commercial applications, the improved mechanical properties are ample justification for the added cost.

34. Cast Steels Containing 0.15 to 0.21 Per Cent Carbon.—Data on the mechanical properties of cast steels containing 0.15 to 0.21 per cent carbon, after the heat treatments shown in Table 10, are given in Table 11. All of these data are from Lorig and Williams,[(545)] but original references are included. It will be noted from the values shown in Table 11 that when the carbon is 0.15 per cent or more the strength is frequently high enough to meet the various specifications for commercial soft-steel castings (60,000 lb. per sq. in., minimum tensile strength), while some of them meet the requirements for the medium grade

TABLE 8.—COMPOSITION AND HEAT TREATMENTS OF CAST STEELS CONTAINING 0.09 TO 0.14 PER CENT CARBON

Number	Composition, per cent					Heat treatment								Original investigator	Reference No.
						Heating				Reheating					
						Temperature		Time, hr.	Cooled in	Temperature		Time, hr.	Cooled in		
	C	Mn	Si	S	P	°C.	°F.			°C.	°F.				
1A	0.09	0.47	0.28	0.01	0.02	855	1575	3	Water	650	1200	2	Air	Giolitti	68
B						1095	2000	14	Water	855	1575	3	Water*		
2A	0.11	0.33	0.27	0.04	0.03	As cast		..	....	...	....	..	....	Hatfield	709
B						900	1650	10	Air						
C						900	1650	10	Air	600	1110	2			
D						900	1650	10	To 600°C.	770	1420	6	Slowly		
E						900	1650	10	Water	600	1110	2			
3A	0.11	0.73	0.27	0.027	0.038	Annealed in commercial furnace					....	..	....	Scharlibbe	138
4A	0.11	0.60	0.40	0.035	0.030	As cast		..	....	...	....	..	....	Oberhoffer	36
B						800	1475	6	Furnace						
C						900	1650	6	Furnace						
D						995	1825	6	Furnace						
5A	0.12	0.32	0.25	0.023	0.011	900	1650	5	Furnace	...	....	..	....	Körber and Pomp	262
6A	0.13	0.43	0.52	0.029	0.035	As cast		..	....	...	....	..	....	Oberhoffer	36
B						800	1475	6	Furnace						
C						890	1635	6	Furnace						
7A	0.13	0.96	0.36	0.067	0.072	As cast		..	....	...	....	..	....	Oberhoffer	36
B						800	1475	6	Furnace						
C						850	1560	6	Furnace						
D						920	1690	6	Furnace						
E						995	1825	6	Furnace						
8A	0.14	0.45	0.29	0.037	0.016	900	1650	5	Furnace	...	....	..	....	Körber and Pomp	262

* Tempered at 650°C. (1200°F.) for 2 hr. and air cooled.

(70,000 lb. per sq. in., minimum tensile strength). In most cases, however, ductility values are too low to meet these specifications.

TABLE 9.—MECHANICAL PROPERTIES OF CAST STEELS CONTAINING 0.09 TO 0.14 PER CENT CARBON

Number	Tensile strength, lb. per sq. in.	Yield strength,* lb. per sq. in.	Elongation, per cent	Gage length	Reduction of area, per cent	Impact value, m-kg. per sq. cm.	Brinell hardness
1*A*	50,500	30,000+	31.8	100 mm.†	55.8		
B	54,500	33,700+	33.0		60.2		
2*A*	42,000	33,000	17	Not given	24		
B	40,000	32,500	38		66		
C	36,500	28,500	38		64		
D	37,000	30,500	39		66		
E	39,500	36,500	38		61.1		
3*A*	56,000	38,000+	33.0	Not given	36.0		
4*A*	59,000	26,000	13.2	$l = 10\,d$	30.0	3.7	126
B	57,000	24,000	28.2		53.0	2.1	119
C	60,000	35,000	29.5		59.5	15.0	116
D	60,000	35,000	31.0		54.0	13.7	126
5*A*	51,000	26,000	36.2	$l = 5\,d$	66.3	4.8‡	
6*A*	42,000	15,000	25.0	$l = 10\,d$	40.9	....	90§
B	42,000	15,000	33.6		64.3	....	90§
C	43,000	26,000	34.5		57.4	....	95§
7*A*	62,000	33,000	13.1	$l = 10\,d$	14.2	2.94	143
B	66,000	28,000	14.4		43.1		
C	68,000	40,000	24.4		40.4	9.39	143
D	68,000	38,000	25.0		38.3	7.09	141
E	69,000	35,000	25.2		45.8	7.42	137
8*A*	59,000	34,000	18.6	$l = 5\,d$	28.7	7.4‡	

* Reported as yield point except values marked + which were reported originally as elastic limit.

† Specimen 13.8 mm. (0.54 in.) in diameter, 100 mm. (3.94 in.) long.

‡ Specimens 30 × 30 × 160 mm. Cylindrical notch 4 mm. in diameter, 15 mm. deep.

§ These Brinell numbers are lower than normal. Oberhoffer's curves for average hardness show that this composition would ordinarily give 115 to 125 Brinell.

To illustrate the properties being regularly obtained in commercial practice, data for five heats of basic open-hearth steel castings are given in Table 12. It is not clear from the text of Fenstermacher's paper[449] how the castings or test coupons were heat treated. Probably, however, they were annealed,

TABLE 10.—COMPOSITION AND HEAT TREATMENT OF CAST STEELS CONTAINING 0.15 TO 0.21 PER CENT CARBON

Number	Composition, per cent					Heat treatment								Original investigator	Reference No.
						Heating			Cooled in	Reheating			Cooled in		
						Temperature		Time, hr.		Temperature		Time, hr.			
	C	Mn	Si	S	P	°C.	°F.			°C.	°F.				
1A	0.15	0.81	0.20	.....	.....	Annealed		..	....	...	....	......	....	Melmoth	272
2A	0.17	0.67	0.23	0.076	0.089	900	1650	5	Furnace	...	....	......	....	Körber and Pomp	262
3A	0.18	0.83	0.30	.....	.....	Annealed		..	....	...	....	......	....	Melmoth	272
4A	0.19	0.62	0.42	0.05	0.035	885	1625	..	Water	95	200	......	....	Grotts	121
B						885	1625	..	Water	425	800				
C						885	1625	..	Water	540	1000				
D						885	1625	..	Water	650	1200				
E						885	1625	..	Water	760	1400				
F						885	1625	..	Water	870	1600				
G						885	1625	..	Water	980	1800				
5A	0.19	0.85	0.32	0.024	0.016	Annealed in commercial furnace					....	......	....	Scharlibbe	138
6A	0.20	0.73	0.40	.....	.....	900	1650	..	Furnace	...	....	......	....	Hall, Nissen, and Taylor	57
7A	0.20	0.72	0.27	0.03	0.04	920	1690	10	Furnace	...	....	......	....	Thum and Holdt	417
8A	0.20	0.46	0.21	0.033	0.045	*	*	..	Kieselguhr	...	....	......	....	Körber and Dryer	72
9A	0.21	0.57	0.21	0.044	0.038	*	*	..	Kieselguhr	...	....	......	....	Körber and Dryer	72
10A	0.21	0.94	0.19	0.03	0.07	As cast		..	....	...	....	......	....	Giolitti	68
B						850	1560	3	Slowly	550	1025				
C						850	1560	..	Air						
D						†									

* Heated 30°C. (55°F.) above critical. [The 29°C. (85°F.) in the Lorig and Williams paper is an error.]
† Forged and heated at 850°C. (1560°F.) for 3 hr., cooled slowly.

normalized, or normalized and annealed in accord with the usual practice when definite minimum mechanical properties are specified.

TABLE 11.—MECHANICAL PROPERTIES OF CAST STEELS CONTAINING 0.15 TO 0.21 PER CENT CARBON

Number	Tensile strength, lb. per sq. in.	Yield strength,* lb. per sq. in.	Elongation, per cent	Gage length	Reduction of area, per cent	Impact value, m-kg. per sq. cm.	Scleroscope hardness
1*A*	62,000	. . . . . .	34.0	2 in.	52.5		
2*A*	64,000	35,000	28.5	$l = 5d$	40.2	3.7†	
3*A*	73,000	. . . . . .	34.0	2 in.	49.0		
4*A*	76,000	47,000	15.0	Not given	26.0	. . . . . .	31.0
B	73,000	46,000	16.0		33.0	. . . . . .	29.5
C	72,000	45,000	17.0		36.0	. . . . . .	29.0
D	71,000	44,000	18.0		39.0	. . . . . .	28.5
E	70,000	42,000	19.0		42.0	. . . . . .	28.0
F	68,000	39,000	21.0		46.0	. . . . . .	27.0
G	66,000	34,000	24.0		51.0	. . . . . .	26.0
5*A*	68,000	43,000+	35.0	Not given	56.0		
6*A*	69,000	43,000+	27.2	Not given	39.7		
7*A*	70,000	47,000	26.9	$l = 5d$	50.9	7.3‡	
8*A*	69,000	41,000	26.9	$l = 10d$	55.7		
9*A*	68,000	45,000	21.0	$l = 10d$	60.5		
10*A*	76,000	43,000+	25.0	2 in.	33.5	4.60†	
B	78,000	53,000+	27.0		42.0	8.13†	
C	80,000	56,000+	28.0		45.0	9.89†	
D	81,000	56,000+	32.0		60.0	17.12†	

* Reported as yield point except values marked + which were reported originally as elastic limit.

† Specimen was 30 × 30 × 160 mm. The cylindrical notch was 4 mm. in diameter and 15 mm. deep.

‡ Specimen was 10 × 10 mm. The notch was 2 mm. wide and 5 mm. deep.

McCrea and Dowdell[392] noted that aluminum-killed acid steels cast in green-sand molds and containing about 0.20 per cent carbon may have higher tensile and yield strengths and lower ductility than required by specifications, presumably due to inclusions. If the normalized specimens are reheated just below the critical—about 705°C. (1300°F.)—the ductility is

improved. Figure 6 shows the results on specimens of representative steels containing 0.18 to 0.21 per cent carbon, 0.55 to 0.74 per cent manganese, and 0.32 to 0.46 per cent silicon, after heating for 2 hr. in a plant furnace at 885 to 900°C. (1625 to 1650°F.) and air cooling, followed by reheating to 705°C. (1300°F.) in a laboratory furnace for the times indicated. The

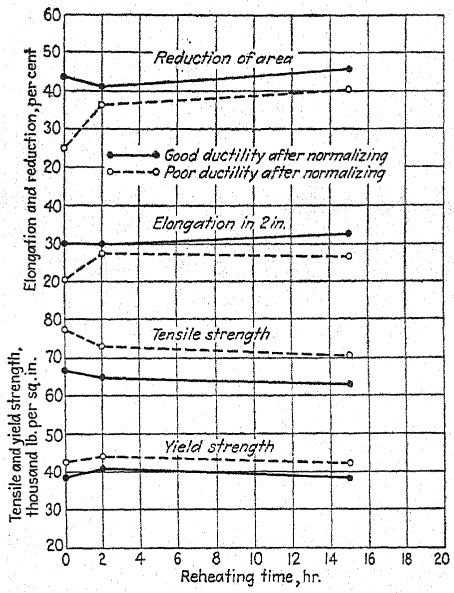

FIG. 6.—Effect of time of reheating at 705°C. (1300°F.) on the tensile properties of aluminum-killed 0.20 per cent carbon cast acid steel. (*McCrea and Dowdell.*[392])

specimens represented by the solid circles had good ductility after the initial normalizing, the others inferior ductility. A 2-hr. tempering treatment at 705°C. (1300°F.) materially increased the ductility of the latter and was found to be the most effective of the treatments tried.

B. MECHANICAL PROPERTIES OF REGULAR-GRADE STEEL CASTINGS

In addition to the properties summarized in Tables 13 to 16, many of which are from experimental heats, considerable information is available on the mechanical properties, as cast and after a variety of heat treatments, of commercial steel castings contain-

ing between 0.22 and 0.35 per cent carbon. This is to be expected as by far the largest commercial production is of castings having a carbon content within this range.

TABLE 12.—MECHANICAL PROPERTIES OF COMMERCIAL LOW-CARBON STEEL CASTINGS*

Number	Composition, per cent					Tensile properties†			
	C	Mn	Si	P	S	Tensile strength, lb. per sq. in.	Yield strength,‡ lb. per sq. in.	Elongation in 2 in., per cent	Reduction of area, per cent
1857	0.19	0.61	0.30	0.029	0.028	68,000	40,500	32.0	55.5
1876	0.21	0.69	0.33	0.018	0.025	73,000	46,500	31.0	50.3
1908	0.16	0.71	0.32	0.018	0.025	75,500	44,000	36.0	62.5
1919	0.21	0.65	0.32	0.025	0.024	76,300	44,900	34.0	54.0
2448	0.20	0.67	0.35	0.020	0.029	73,000	41,000	32.0	57.7
Average	0.19	0.67	0.30	0.022	0.026	73,100	43,500	33.0	56.0

* Fenstermacher.(449)

† Tests on specimens from 1 × 2 × 10-in. lugs removed from castings after heat treating.

‡ Reported as elastic limit.

35. Lorig and Williams' Data.—The mechanical properties of regular-grade cast steels as summarized by Lorig and Williams(545) are given in Tables 13 to 16. The results by Mitchell(393) have been deleted, as these show clearly the effect of heat treatment on commercial castings and so are given in detail on page 65. Values for a 0.29 per cent carbon steel, from Hatfield,(709) have been added to Table 15. Tables 13 and 14 give the compositions, heat treatments, and properties of steels containing 0.23 to 0.28 per cent carbon; Tables 15 and 16 give the corresponding data on steels containing 0.29 to 0.34 per cent carbon.

The data given in Tables 14 and 16 show that in general the strength and hardness increase and the ductility decreases as the carbon content increases.

36. Properties of Regular-grade Castings Obtainable in Commercial Production.—Several investigators have reported data on castings made in regular production. A few of these data by Fenstermacher,(449) Gregg,(617) and Mitchell(393) are given in Table 17.

TABLE 13.—COMPOSITION AND HEAT TREATMENT OF CAST STEELS CONTAINING 0.23 TO 0.28 PER CENT CARBON

Number	Composition, per cent					Heat treatment								Original investigator	Reference No.
	C	Mn	Si	S	P	Heating			Cooled in	Reheating			Cooled in		
						Temperature		Time, hr.		Temperature		Time, hr.			
						°C.	°F.			°C.	°F.				
1A	0.23	0.66	0.21	0.03	0.05	875	1610	..	36 hr.	...	....	..	....	Giolitti	68
B						825	1520	12	10 hr.						
C						875	1610	..	Slowly	800	1475	⅓	Hot water*		
2A	0.24	0.78	0.28	.....	.....	900	1650	..	Furnace	...	....	..	....	Hall, Nissen, and Taylor	57
3A	0.25	0.68	0.32	0.032	0.012	As received		..	....	...	....	..	....	Moore	191
B						900	1650	1	Air	830	1525	1	Air		
C						900	1650	1	Air	830	1525	1	Air*		
D						900	1650	1	Air	830	1525	1	Air*		
4A	0.25	0.44	0.29	0.07	0.105	Bessemer steel, as cast					....	..	....	Körber and Pomp	161
B						Same, annealed, slowly cooled									
C						Same, annealed, rapidly cooled									
5A	0.25	0.71	0.13	0.030	0.021	Basic open-hearth steel, as cast					....	..	....	Körber and Pomp	161
B						Same, annealed, slowly cooled									
C						Same, annealed, rapidly cooled									
6A	0.26	0.84	0.37	.....	.....	Annealed		..	Furnace	...	....	..	....	Melmoth	272
7A	0.27	0.71	0.41	.....	.....	900	1650	..	Water	705	1300	..	Furnace	Hall, Nissen, and Taylor	57
8A	0.28	0.74	0.37	0.028	0.036	920	1690	..	Furnace	...	....	..	....	Körber and Pomp	262
9A	0.28	0.65	0.27	0.032	0.027	840	1550	7	†	...	....	..	....	Bull	443

* Also tempered:
Steel 1C, 1 hr. at 650°C. (1200°F.).
Steel 3C, 1 hr. at 315°C. (600°F.) air cooled.
Steel 3D, 1 hr. at 540°C. (1000°F.) air cooled.
Furnace cooled to 540°C. (1000°F.), then cooled in air.

The results by Fenstermacher were obtained on basic-open hearth castings over a period of one month. The specimens were machined from 1 × 2 × 10-in. lugs cut from the castings which were presumably heat treated. Details are not given;

TABLE 14.—MECHANICAL PROPERTIES OF CAST STEELS CONTAINING 0.23 TO 0.28 PER CENT CARBON

Number	Tensile strength, lb. per sq. in.	Yield strength,* lb. per sq. in.	Elongation, per cent	Gage length	Reduction of area, per cent	Impact value		Brinell hardness
						Charpy, ft-lb.	m-kg. per sq. cm.	
1*A*	62,000	34,000^{+}	13.0	Not given				
B	68,000	41,000^{+}	17.0		13.9			
C	73,000	47,000^{+}	17.5		24.9			
2*A*	71,000	42,000^{+}	28.6	Not given	47.8			
3*A*	67,000	27,000	22.0	2 in.	33.0	20.1	.....	119
B	77,000	44,000	30.5		51.0	32.6	.....	136
C	77,000	43,000	31.5		52.0	32.0	.....	136
D	76,000	43,000	31.7		56.0	34.0	.....	133
4*A*	56,000	......	5.2	$l = 5d$	4.5	....	1.4†	
B	64,000	......	15.9		16.5	....	1.1†	
C	70,000	......	8.3		12.8	....	2.9†	
5*A*	56,000	......	22.6	$l = 5d$	37.3	....	2.1†	
B	68,000	......	27.8		51.3	....	2.0†	
C	74,000	......	20.3		30.3	....	8.5†	
6*A*	75,000	......	33.0	2 in.	54.2			
7*A*	72,000	46,000^{+}	32.9	Not given	57.6	....	35.5‡	
8*A*	75,000	36,000	18.3	$l = 5d$	28.3	....	1.5†	
9*A*	74,000	43,000	28.0	2 in.	42.0			

* Reported as yield point except values marked $^{+}$ which were originally reported as elastic limit.

† Specimens were 30 × 30 × 160 mm. The cylindrical notch was 4 mm. in diameter, 15 mm. deep.

‡ Frémont bar; value given in m-kg.

probably the heat treatment consisted of annealing, normalizing, or both.

Gregg's[617] data are from tests on lugs cast on the rim of locomotive-wheel centers which in the rough had an over-all diameter of 56 in. and weighed 1650 lb. The castings were water

TABLE 15.—COMPOSITION AND HEAT TREATMENT OF CAST STEELS CONTAINING 0.29 TO 0.34 PER CENT CARBON

Number	Composition, per cent					Heat treatment								Original investigator	Reference No.
						Heating				Reheating					
	C	Mn	Si	S	P	Temperature °C.	Temperature °F.	Time, hr.	Cooled in	Temperature °C.	Temperature °F.	Time, hr.	Cooled in		
1*A*	0.29	0.81	0.24	0.03	0.03	As cast		. .	. . . .	. . .	. . . .	. .	. . . .	Hatfield	709
B						900	1650	6	Air						
C						880	1615	20	Air						
D						850	1560	8	To 610°C.	610	1130	2	Slowly		
E						900	1650	6	Air	630	1165	2	Furnace		
F						900	1650	6	To 610°C.	610	1130	2	Sand		
G						880	1615	20	To 600°C.	770	1420	6	Air		
2*A*	0.30	0.62	0.18	0.038	0.038	900	1650	2	Water	650	1200	. .	Furnace	Langenberg	74
B						900	1650	2	Furnace						
C						950	1740	2	Air	500	930	2	Furnace		
3*A*	0.31	0.94	0.31	. . . . .	. . . .	900	1650	. .	Furnace	. . .	. . . .	. .	. . . .	Hall, Nissen, and Taylor	57
4*A*	0.34	0.84	0.39	0.09	0.031	As cast		. .	. . . .	. . .	. . . .	. .	. . . .	Körber and Pomp	161
B						Annealed		. .	Slowly						
C						Annealed		. .	Rapidly						
5*A*	0.34	0.96	0.19	0.02	0.06	As cast		. .	. . . .	. . .	. . . .	. .	. . . .	Giolitti	68
B						850	1560	. .	Slowly						
C						850	1560	. .	Air	550	1020				
D						*									
6*A*	0.34	0.86	0.61	. . . . .	. . . . .	As received (treatment not known)					. . . .	. .	. . . .	Hall, Nissen, and Taylor	57
B						850	1560	4	Water	675	1250	6	Air		
C						850	1560	4	Air	700	1290	6	Air		

* Forged and heated at 850°C. (1560°F.) for 3 hr., cooled slowly.

quenched after holding 3 hr. at 900°C. (1650°F.), tempered for 4.5 hr. at 620°C. (1150°F.), and furnace cooled. In view of the opinion expressed by some that high manganese is accompanied by brittleness, the high elongation and especially the

TABLE 16.—MECHANICAL PROPERTIES OF CAST STEELS CONTAINING 0.29 TO 0.34 PER CENT CARBON

Number	Tensile strength, lb. per sq. in.	Yield strength,* lb. per sq. in.	Elongation, per cent	Gage length	Reduction of area, per cent	Impact values: Charpy, ft-lb.	Impact values: m-kg. per sq. cm.	Brinell hardness
1*A*	79,000	44,000	10	Not given	12.0			
B	84,000	52,500	20		27.6			
C	79,000	49,500	26		30.8			
D	76,000	48,500	17		21.6			
E	83,500	49,500	23		30.8			
F	74,500	39,500	30		44.8			
G	78,500	49,000	25		36.3			
2*A*	87,000	57,000	15.3	Not given	22.8	14.25	......	176
B	79,000	46,000	18.3		27.2	11.32	......	156
C	83,000	50,000	16.2		23.9	11.87	......	159
3*A*	78,000	48,000+	26.2	Not given	41.3			
4*A*	85,000		10.0	$l = 5d$	10.0	.....	1.1†	
B	79,000	40,000	18.9		24.7	.....	3.3†	
C	86,000	54,000	14.3		19.3	.....	4.8†	
5*A*	87,000	51,000+	18.0	2 in.	21.0	.....	1.7†	
B	86,000	55,000+	25.0		33.5	.....	6.15†	
C	91,000	63,000+	24.0		40.5	.....	7.41†	
D	89,000	59,000+	30.0		53.0	.....	11.75†	
6*A*	88,000	55,000+	24.3	Not given	35.6	.....	32.5‡	
B	83,000	55,000+	21.0		32.7	.....	30.0‡	
C	97,000	58,000+	22.2		31.4	.....	12.5‡	

* Reported as yield point except values marked + which were reported originally as elastic limit.

† Specimens were 30 × 30 × 160 mm. The cylindrical notch was 4 mm. in diameter, 15 mm. deep.

‡ Frémont bar; values given in m-kg.

high value for reduction of area shown in Gregg's data in Table 17 are significant. Another interesting feature in connection with Gregg's work is that these rather complicated castings were given a water quench, and no trouble was encountered from cracking.

TABLE 17.—AVERAGE MECHANICAL PROPERTIES OF REGULAR-GRADE COMMERCIAL CASTINGS*

Composition and properties	Fenster-macher,[449] average of 54 tests	Gregg,[617] average of 94 tests	Mitchell,[393] average of 80 tests	
Carbon, per cent	0.22 to 0.30	0.28 to 0.35	0.30†	
Manganese, per cent	0.65 to 0.80	0.85 to 0.90	0.79	
Silicon, per cent	0.30 to 0.35	0.35 to 0.45	0.30	
Sulphur, per cent	0.020 to 0.030	About 0.030	0.026	
Phosphorus, per cent	0.015 to 0.035	About 0.037	0.030	
			(a)	(b)
Tensile strength, lb. per sq. in.	76,000	94,000	74,100	75,000
Yield strength, lb. per sq. in.	44,200	69,500	37,100	41,500
Elongation in 2 in., per cent	30.3	26.0	19.5	24.5
Reduction of area, per cent	47.6	61.0	31.0	46.2
Brinell hardness	...	180	160	145
Izod impact, ft-lb	...	40 to 45	16	20
Endurance ratio	...	...	0.40	0.44

* See text, section 36, for details of heat treatment.

† Average of 80 specimens of basic open-hearth and acid electric steels. Analyses very uniform.

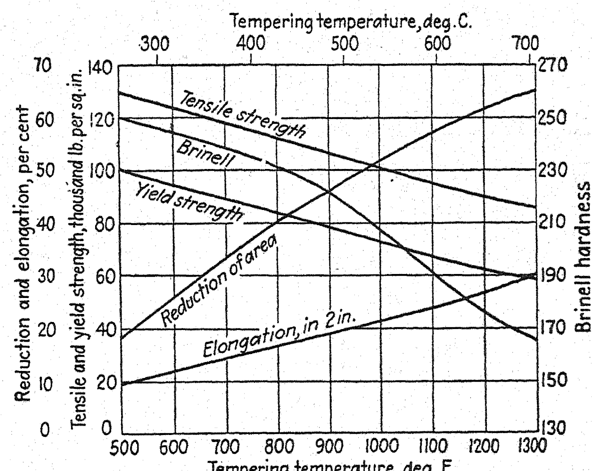

FIG. 7.—Effect of tempering on mechanical properties of 1.25-in. sq. cast bars (see Table 17 for composition), water quenched from 900°C. (1650°F.). (*Mitchell.*[393])

The data from Mitchell's paper[393] reproduced in Table 17 are average properties determined on: (*a*) specimens cast in 1.25-in. sq. bars, and (*b*) specimens annealed at 900°C. (1650°F.)

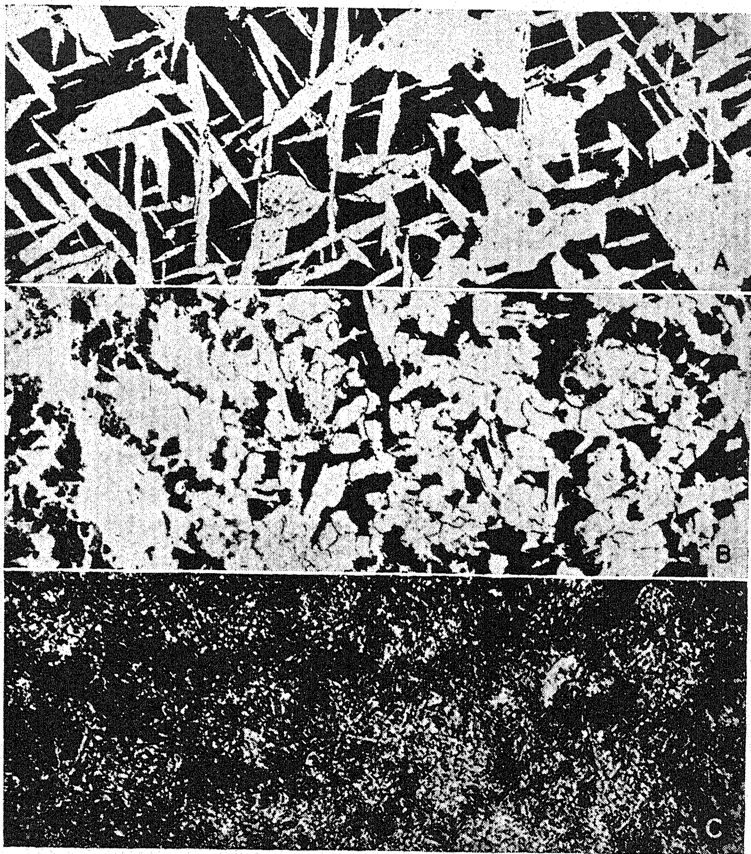

FIG. 8.—Structure of 0.30 per cent carbon, 0.79 per cent manganese, cast steel. *A* = as cast; *B* = annealed; *C* = water quenched and tempered at 705°C. (1300°F.). 100 × (*Courtesy of Bonney-Floyd Company.*)

and cooled slowly. He also determined the properties of the material after quenching in water from 900°C. (1650°F.) and tempering at 260, 370, 480, 595, and 705°C. (500, 700, 900, 1100, and 1300°F.). These data are given in Fig. 7. Typical microstructures of specimens as cast, as annealed, and as quenched and tempered at 705°C. (1300°F.) are shown in Fig. 8.

Mitchell also calculated the average improvement in mechanical properties of cast carbon steel of the composition given in Table 17 by (1) annealing and (2) water quenching from 900°C. (1650°F.) and tempering at 705°C. (1300°F.). The percentage increase was based upon the properties of 1.25-in. sq. cast bars as unity:

Property	Increase, per cent	
	Annealed specimens	Quenched and tempered specimens
Tensile strength	1.1	12.6
Yield strength	12.0	53.8
Elongation in 2 in	30.8	61.5
Reduction of area	51.7	124.1
Brinell hardness	8.3*	2.5
Izod impact	23.5	158.8
Endurance ratio	10.0	25.5

* Decrease.

The importance of the impact properties of regular-grade steel castings was emphasized by Grotts,[619] who maintained that lack of ability to resist impact has been the cause of many so-called mysterious failures encountered in service. He found no relation between tensile strength and impact resistance but observed an increase in impact resistance with increase in yield strength. He concluded that metallographic examination gave the best indication of the probable impact properties and that these properties are markedly influenced by heat treatment.

For use as a base line in studying the mechanical properties of cast alloy steels, Armstrong[790] determined properties on a 0.5-ton heat of basic electric carbon steel containing 0.30 per cent carbon, 0.70 per cent manganese, and 0.28 per cent silicon. The work was done at Norfolk Navy Yard as part of an investigation to determine the best chemical composition and heat treatment for ship castings. The data on the cast carbon steel are shown in Table 18. Armstrong found, for castings which could not readily be subjected to quenching and tempering, that normaliz-

ing followed by tempering resulted in better yield strength and impact values than any of the annealing treatments used.

From the foregoing discussion it is evident that 0.22 to 0.35 per cent carbon steel can be cast and treated to produce a wide range of mechanical properties, thus making this grade suitable for practically all the requirements of the industries using carbon-steel castings. There are some applications for which low-carbon steel castings are preferred and, as indicated later, other applications for which high-carbon castings are most suitable; for the major tonnage of castings, however, present practice tends toward using cast steel within the regular-grade carbon range, modifying the properties by heat treatment or alloy additions as required.

C. MECHANICAL PROPERTIES OF HIGH-CARBON STEEL CASTINGS

There have been a number of reports of mechanical properties of cast carbon steel containing between 0.36 and 0.60 per cent carbon, but only two, by Oberhoffer in Germany and Arnold in England, on steels containing more than 0.60 per cent.

37. Cast Steels Containing 0.36 to 0.60 Per Cent Carbon.—Lorig and Williams(545) have also summarized existing data on the properties of high-carbon steel castings. Part of their summary is given in Tables 19 and 20. The tensile strength after heat treatment varies from about 80,000 lb. per sq. in. for the lower carbon contents to approximately 90,000 lb. per sq. in. for the 0.46 to 0.53 per cent carbon castings. Ductility as represented by the elongation is in most cases satisfactory, ranging from 16 to 27 per cent, depending upon the carbon content and heat treatment. The highest values were obtained by the heat treatments given to specimens 1*B*, 2*A*, 6*B*, 6*C*, 8*B*, and 9*B*.

The effect of tempering after quenching cast steels containing 0.19 per cent carbon and 0.62 per cent manganese, and 0.38 per cent carbon and 0.61 per cent manganese (sulphur and phosphorus 0.05 per cent or less) was studied by Grotts.(121) Values for the low-carbon material are included in Tables 10 and 11, pages 57 and 58. Values for the higher carbon grade are given in Table 21.

TABLE 18.—ARMSTRONG'S[790] DATA ON CAST CARBON STEEL, CONTAINING 0.30 PER CENT CARBON, 0.70 PER CENT MANGANESE, AND 0.28 PER CENT SILICON

Number	Tensile strength, lb. per sq. in.	Yield strength,* lb. per sq. in.	Elongation,† per cent	Reduction of area, per cent	Izod impact value, ft-lb.	Treatment
1*A*	81,400	41,600	22.0	34.5	8.0	As cast
1*B*	78,700	44,250	28.0	45.0	20.5	Annealed at 900°C. (1650°F.)
1*C*	77,000	43,400	29.0	45.0	......	Annealed at 925°C. (1700°F.)
1*D*	77,250	46,750	29.5	48.0	......	Double annealed at 900°C. (1650°F.)
1*E*	77,200	45,000	29.5	47.0	22 to 28	Double annealed at 900 and 830°C. (1650 and 1525°F.)
1*F*	76,000	40,750	30.5	50.0	......	Double annealed at 925 and 815°C. (1700 and 1500°F.)
1*G*	79,000	50,700	30.0	47.0	......	Normalized at 925°C. (1700°F.). Annealed at 830°C. (1525°F.)
1*H*	79,000	45,000	31.0	47.0	34.0	Normalized at 900°C. (1650°F.). Tempered at 650°C. (1200°F.)
1*J*	79,600	49,400	32.0	53.0	41.0	Double normalized at 900°C. (1650°F.). Tempered at 650°C. (1200°F.)
1*K*	76,000	44,500	31.0	56.0	33.0	Double normalized at 925 and 830°C. (1700 and 1525°F.). Tempered at 705°C. (1300°F.)
1*L*	78,250	45,250	32.0	53.5	27.0	Normalized at 900°C. (1650°F.). Tempered at 720°C. (1325°F.)
1*M*	90,750	60,500	27.0	57.0	46.0	Water quenched from 900°C. (1650°F.). Tempered at 675°C. (1250°F.)
1*N*	74,000	48,750	30.0	58.5	37.0	Water quenched at 900°C. (1650°F.). Tempered at 720°C. (1325°F.)

* Reported as yield point.
† Gage length not given, probably 2 in.

The effect of varying the annealing temperature was investigated by Oberhoffer[36] for four high-carbon cast steels* of the following analysis:

Steel	Composition, per cent				
	C	Mn	Si	S	P
A	0.46	0.92	0.20	0.042	0.041
B	0.53	0.79	0.25	0.036	0.027
C	0.69	1.03	0.25	0.022	0.021
D	0.86	0.90	0.27	0.028	0.016

* Oberhoffer's paper also included results on a low-carbon steel, annealed at increasing temperatures. These are given in Tables 8 and 9, pp. 55 and 56.

TABLE 19.—COMPOSITION AND HEAT TREATMENT OF CAST STEELS CONTAINING 0.36 TO 0.60 PER CENT CARBON

Number	Composition, per cent					Heat treatment								Original investigator	Reference No.
						Heating				Reheating					
	C	Mn	Si	S	P	Temperature °C.	Temperature °F.	Time, hr.	Cooled in	Temperature °C.	Temperature °F.	Time, hr.	Cooled in		
1A	0.37	0.79	0.40	0.008	0.019	As received		....	....	...	....	..	....	Hall, Nissen, and Taylor	57
B						900	1650	4	Water	680		6	Air		
C						900	1650	4	Air	695	1260	6	Air		
											1290				
2A	0.39	0.73	0.44	.....	.....	900	1650	4.5	Air	695		6	Furnace	Hall, Nissen, and Taylor	57
											1290				
3A	0.39	0.86	0.41	0.008	0.019	As received		....	....	...		..	....	Hall, Nissen, and Taylor	57
B						900	1650	4	Water	680	....	6	Air		
C						900	1650	4	Air	695	1260	6	Air		
											1290				
4A	0.40	0.63	0.30	0.069	0.058	950	1740	2	Air	500		4	....	Langenberg	74
											930				
5A	0.42	0.60	0.27	0.066	0.068	950	1740	2	Air	500		4	....	Langenberg	74
B	0.42	0.65	0.28	0.069	0.043	950	1740	2	Air	500	930	4			
											930				
6A	0.42	0.69	0.43	.....	.....	900	1650	4	Furnace	...		..	....	Hall, Nissen, and Taylor	57
B	0.42	0.71	0.54	.....	.....	900	1650	4	Oil	675	....	6	Furnace		
C						900	1650	4	Water	675	1250	6	Furnace		
											1250				
7A	0.46	0.73	0.28	.....	.....	Annealed, cooled in furnace						..	....	Melmoth	272
											....				
8A	0.47	0.63	0.27	0.029	0.03	750	1380	12	Slowly	...		..	....	Giolitti	68
B						750	1380	12	Slowly	850	....	7	Air		
											1560				
9A	0.48	0.68	0.41	0.01	0.019	As received		....	....	...		..	....	Hall, Nissen, and Taylor	57
B						900	1650	4	Water	675	....	6	Air		
C						900	1650	4	Air	695	1250	6	Air		
											1290				
10A	0.51	0.56	0.38	.....	.....	900	1650	4	Air	760		6	Furnace	Hall, Nissen, and Taylor	57
B	0.51	0.69	0.44	.....	.....	920	1690	5	Air	695	1400	6	Furnace		
											1290				
11A	0.53	0.28	0.22	0.046	0.058	940	1725	1.25	Rapidly to	750	1380	..	Slowly	Körber and Pomp	262

The manganese in specimens *A*, *C*, and *D* is so high that it would be more accurate to class these as manganese steels. The mechanical properties as cast and as influenced by the

TABLE 20.—MECHANICAL PROPERTIES OF CAST STEELS, CONTAINING 0.36 TO 0.60 PER CENT CARBON

Number	Tensile strength, lb. per sq. in.	Yield strength,* lb. per sq. in.	Elongation, per cent	Gage length	Reduction of area, per cent	Impact value			Brinell hardness
						Charpy, ft-lb.	m-kg. per sq. cm.	Frémont, m-kg.	
1*A*	84,000	40,000	23.9	Not given	32.9	...	....	5	
B	82,000	50,000	26.7		49.9	...	....	10	
C	88,000	49,000	21.4		28.3	...	....	6	
2*A*	79,000	41,000	26.2	Not given	35.7	...	....	8.8	
3*A*	72,000	35,000	16.8	Not given	31.4	...	....	6.0	
B	86,000	51,000	23.5		38.7	...	....	24.5	
C	83,000	43,000	20.7		29.5	...	....	14.0	
4*A*	87,000	51,000	17.0	Not given	20.5	4.6	....	....	182
5*A*	89,000	52,000	19.2	Not given	20.4	3.5			
B	88,000	52,000	18.2		20.4	5.5	....	....	156
6*A*	77,000	40,000	22.0	Not given	25.0	...	....	6.5	
B	81,000	50,000	23.9		37.9	...	....	20.5	
C	82,000	51,000	26.4		44.2	...	....	17.7	
7*A*	93,000	......	22.0	2 in.	33.6				
8*A*	76,000	......	19.0	†	....	...	1.2†		
B	88,000	......	24.0		....	...	3.6†		
9*A*	83,000	39,000	22.6	Not given	27.1	...	....	7	
B	88,000	52,000	24.9		41.9	...	....	10	
C	91,000	54,000	21.3		29.5	...	....	8.5	
10*A*	83,000	42,000	19.5	Not given	19.2	...	....	5.8	
B	84,000	47,000	22.5		26.6	...	....	5.9	
11*A*	81,000	28,000	16.1	$l = 5d$	15.9	...	1.3		

* Reported as elastic limit.

† The tensile specimens were 13.8 mm. (0.54 in.) in diameter, 100 mm. (3.94 in.) long. The impact specimens were 30 × 30 × 160 mm.; with a cylindrical notch 4 mm. in diameter and 15 mm. deep.

annealing temperatures are given in Table 22. The effect of increasing the annealing temperature seems to be slight, after the critical temperature is exceeded.

As the ductility of steel castings containing 0.36 to 0.60 per cent carbon is frequently low as cast, there is no apparent advantage in specifying this grade unless the mechanical properties are to be improved by heat treatment. The demand for this grade of castings is generally limited to applications where high strength and hardness or wear resistance are required, such as in rolls, dies, machine tools, and some classes of railroad and automotive equipment.

TABLE 21.—EFFECT OF TEMPERING ON THE MECHANICAL PROPERTIES OF A 0.38 PER CENT CAST CARBON STEEL*

Water quenched and tempered at		Tensile strength, lb. per sq. in.	Yield strength,† lb. per sq. in.	Elongation,‡ per cent	Reduction of area, per cent	Scleroscope hardness
°C.	°F.					
95	200	140,000	140,000	0	0	50
425	800	122,000	118,000	5.0	8.0	46
540	1000	113,000	100,000	7.5	12.0	42
650	1200	102,000	82,000	10.0	15.0	37
760	1400	98,000	73,000	14.0	21.0	33
870	1600	92,000	70,000	19.0	28.0	30

* Grotts.[121]
† Reported as yield point.
‡ Gage length not given, probably 2 in.

38. High-carbon Centrifugal Castings.—Forgings for big guns must be made of extremely clean steel which has, consequently, satisfactory impact resistance on radial specimens. To fulfill this requirement, and to reduce cost and increase production, the centrifugal casting of gun tubes, especially those to be slightly cold worked by "autofrettage" (enlarging the diameter and decreasing the length by internal hydraulic pressure), has been developed at Watertown Arsenal.

Alloy steels are normally used; there are only a few data on plain carbon steel. Dickson[363] furnished information on one containing 0.40 per cent carbon, 0.70 per cent manganese, 0.26 per cent silicon, 0.03 per cent sulphur, and 0.024 per cent phosphorus After a complex heat treatment, and after cold working to an average bore enlargement of 6.7 per cent followed by annealing, the average properties were as shown on page 74.

39. Extra-high-carbon Cast Steels (Carbon over 0.60 Per Cent).—Information available on mechanical properties of cast steels containing more than 0.60 per cent carbon is limited. Two

TABLE 22.—EFFECT OF ANNEALING TEMPERATURES ON MECHANICAL PROPERTIES* OF FOUR CAST STEELS†

Number	Carbon, per cent	Furnace cooled after 6 hr. at		Tensile strength, lb. per sq. in.	Yield strength, lb. per sq. in.	Elongation,‡ per cent	Reduction of area, per cent	Impact value, m-kg. per sq. cm.	Brinell hardness
		°C.	°F.						
A1	0.46	As cast	....	79,200	40,300	3.0	3.83		
2		750	1380	84,800	40,100	7.07	8.00		
3		775	1425	89,500	49,900	7.50	8.73		
4		800	1470	80,000	48,200	7.57	8.43		
5		850	1560	87,200	47,600	6.17	7.00		
6		900	1650	88,800	46,200	7.07	8.30		
B1	0.53	As cast	....	87,900	34,800	6.47	4.07	1.26	213
2		700	1290	79,200	34,800	5.07	4.60	1.29	200
3		760	1400	90,700	38,700	11.03	17.87	1.40	194
4		790	1455	98,100	49,300	13.97	18.13	2.52	198
5		820	1510	99,300	59,900	15.93	18.03	3.53	208
6		850	1560	98,700	49,200	14.50	16.90	1.76	212
7		880	1615	95,000	47,600	12.45	14.00	2.64	209
8		910	1670	100,600	45,800	15.83	20.50	1.67	211
9		950	1740	100,300	45,400	13.30	15.70	1.42	219
10		1000	1830	100,400	46,200	13.80	15.57	1.49	215
C1	0.69	As cast	....	88,900	52,800	3.77	1.33	1.45	245
2		700	1290	87,300	38,500	7.03	4.70	1.34	196
3		730	1345	108,100	46,900	7.47	7.93	1.40	223
4		760	1400	110,800	47,200	6.63	7.90	1.25	235
5		790	1455	109,800	48,200	9.63	9.33	1.15	223
6		820	1510	109,700	47,100	8.10	7.80	1.68	235
7		1000	1830	106,700	48,800	4.93	3.63	1.48	243
D1	0.86	As cast	....	71,300	43,200	1.40	0.37	1.35	255
2		650	1200	89,900	44,700	3.23	2.90	1.23	241
3		680	1255	88,300	41,200	2.70	2.07	1.45	239
4		710	1310	107,500	45,900	8.63	6.47	2.03	215
5		740	1365	109,200	48,200	4.90	5.03	1.33	236
6		800	1470	108,000	50,500	4.06	3.58	1.43	253
7		1000	1830	116,900	57,500	3.40	1.80	1.32	271

* Each value is the average of three determinations.
† Oberhoffer.(36)
‡ Specimens were 200 mm. (7.875 in.) long and 20 mm. (0.787 in.) in diameter.

of Oberhoffer's steels, properties of which are given in Table 22, contained 0.69 and 0.86 per cent carbon respectively. The data indicate that both of these steels are strong and brittle,

and that increasing the annealing temperature has little effect on the properties after the critical range is exceeded. The specimens containing 0.69 per cent carbon are somewhat more ductile than those containing 0.86 per cent carbon.

The properties of a series of cast steels containing between 0.70 and 1.95 per cent carbon were determined by Arnold.[4] Tests were made on specimens as cast and as annealed for 70 hr.

Property	Chill cast and annealed at 950°C. (1740°F.), air cooled; reannealed at 850°C. (1560°F.), furnace cooled; water quenched from 825°C. (1520°F.), and tempered at 625°C. (1160°F.)	Cold worked to a bore enlargement of 6.7 per cent and annealed at 300°C. (570°F.)
Tensile strength, lb. per sq. in.	100,000 to 108,000	104,000 to 116,000
Yield strength, lb. per sq. in.	62,000 to 70,000	88,000 to 96,000
Elongation in 2 in., per cent	14 to 16	5 to 11
Reduction of area, per cent	25 to 55	36 to 48

at 950°C. (1740°F.). The results are given in Table 23. Despite packing the samples in lime there was considerable carbon loss in annealing, varying from 6 points for the 0.86 per cent carbon steel to 85 points for the 1.95 per cent carbon sample (see analyses marked *B* in Table 23).

Although the values for these very high carbon steels are erratic, there is undeniable evidence from both Arnold's and Oberhoffer's data that the ductility of castings in this carbon range is extremely low and may be zero. In some specimens annealing improved the ductility, in others no improvement could be noted. Some further consideration is given to cast steels of this grade in the section on compressive properties.

The possible uses of extra-high-carbon steel castings appear to be extremely limited, especially as their properties as cast are not much superior to those of a high-grade cast iron, and after heat

treatment their properties are decidedly inferior to those of heat-treated wrought steels of approximately the same analysis.

TABLE 23.—MECHANICAL PROPERTIES OF EXTRA-HIGH-CARBON CAST STEELS*

Number	Composition, per cent					Condition	Tensile strength, lb. per sq. in.	Yield strength, lb. per sq. in.	Elongation in 2 in., per cent	Reduction of area, per cent
	C	Mn	Si	S	P					
1*A*	0.70	0.11	0.10	0.045	.....	Cast	45,100	41,500	1.5	1.8
B	0.72					Annealed	68,000	31,200	6.0	5.3
2*A*	0.86	0.06	0.05	0.02	0.02	Cast	59,400	59,400	1.5	2.1
B	0.80					Annealed	53,300	34,700	2.5	4.2
3*A*	0.97	0.06	0.03	0.025	0.018	Cast	72,600	49,900	2.0	1.8
B	0.83					Annealed	65,100	41,500	4.0	1.7
4*A*	1.29	0.10	0.28	0.02	0.02	Cast	49,900	49,900	0	0
B	1.10					Annealed	67,000	37,100	2.5	3.5
7*A*	1.95	0.03	0.02	0.015	0.016	Cast	29,800	29,800	0	0
B	1.10					Annealed	28,800	28,800	0	0
8*A*	1.76	0.06	0.07	0.020	0.022	Cast	50,000	50,000	0	0
B	1.38					Annealed	31,600	26,200	0	0

* Arnold.(4)

D. TORSIONAL AND COMPRESSIVE PROPERTIES

There are few data on the torsional and compressive properties of cast carbon steel. Lessells made torsion tests on a 0.33 per cent carbon steel, and Hatfield on a 0.25 per cent carbon steel. These results, and the data by Arnold on compressive properties, are the only ones available.

40. Torsional Properties of Steel Castings.—Steel castings may be subjected in practice to torsional loads, although torsional properties are not included in specifications and are rarely mentioned in the literature. Lessells(127) reported on the torsional and other properties of cast-steel specimens in the cast, annealed, and normalized conditions. The analysis of the steel was 0.33 per cent carbon, 0.77 per cent manganese, 0.32 per cent silicon, 0.04 per cent phosphorus, and 0.036 per cent sulphur. The specimens were coupons or lugs ($1\frac{5}{16} \times 2\frac{1}{2} \times 7\frac{1}{4}$ in.) on large castings and, after removal from the castings, were

annealed or normalized at 925°C. (1700°F.). The results are shown in Table 24. Lessells' data indicate that the ultimate strength in torsion is lower than in tension, but this relation

TABLE 24.—TORSIONAL AND OTHER PROPERTIES OF STEEL CASTINGS CONTAINING 0.33 PER CENT CARBON AND 0.77 PER CENT MANGANESE*

Condition	Tensile properties				
	Tensile strength, lb. per sq. in.	Yield strength, lb. per sq. in.	Elastic limit, lb. per sq. in.	Elongation in 2 in., per cent	Reduction of area, per cent
Cast	76,200	33,500	19,900	26.0	34
Annealed	79,800	41,200	37,100	26.8	39.5
Normalized	84,900	45,800	40,500	27.6	45.6

Condition	Torsional properties			
	Ultimate stress, lb. per sq. in.	Yield point, lb. per sq. in.	Elastic limit, lb. per sq. in.	Deg. twist
Cast	60,500	34,600	24,500	460
Annealed	59,500	25,800	25,600	840
Normalized	62,600	28,500	24,300	1025

Condition	Shock and hardness properties		
	Repeated blow	Single blow	Brinell hardness
Cast	676 blows	5.8 ft-lb.	141
Annealed	980 blows	13.0 ft-lb.	143
Normalized	2000 blows	18.0 ft-lb.	163

* Lessells.(127)

did not always hold for the elastic limit and yield strength. The degree of twist showed a better relation to the shock properties than it did to the elongation and reduction of area.

In discussing the most suitable steels for automobile parts in 1920, Hatfield[58] reported the torsional properties of well-annealed steel castings of 0.20 to 0.25 per cent carbon content, shown in Table 25.

TABLE 25.—TORSIONAL AND OTHER PROPERTIES OF STEEL CASTINGS CONTAINING 0.20 TO 0.25 PER CENT CARBON*

Tensile Properties	
Tensile strength, lb. per sq. in.	70,400
Yield strength, lb. per sq. in.	35,800
Elastic limit, lb. per sq. in.	31,600
Elongation, per cent	35.5
Reduction of area, per cent	53.4
Torsional Properties	
Maximum stress,† lb. per sq. in.	49,000
Elastic limit, lb. per sq. in.	20,200
Angle of twist, deg	700
Hardness and Impact Properties	
Brinell hardness	127
Charpy impact, ft-lb	9.0
Izod impact, ft-lb	19.0

* Hatfield.[58]

† Given as probable value.

From the data available it is not possible to generalize on the relation between the tensile and torsional properties of steel castings.

41. Compressive Properties of Steel Castings.—Steel castings are rarely, if ever, used in engineering service where the only requirement is resistance to compression. There are two reasons for this; in the first place, there are few uses where resistance to compression is the only requirement, and in the second place, cast iron is a cheaper material and has a relatively high strength in simple compression. In both wrought and cast steels, it may be assumed that the modulus of elasticity, elastic limit, and yield strength are essentially the same in compression as in tension. The published information on the compressive properties of steel castings is extremely meager. Arnold[4], in 1901, published the results of a series of compression tests on cast crucible steels of different carbon contents. His tests were made on cylinders 0.564 in. in diameter and 2 in. in height, and he determined the percentage decrease in height under a load of 100 tons (224,000 lb.) per sq. in. using both cast and annealed specimens. Because of the erratic character of his tensile-test data (see Table

23) no consistent relation could be found between tensile strength and compression in the various tests. The correlation between carbon content and compression is fairly good, as shown in Fig. 9. The percentage compression under a load of 100 tons (224,000 lb.) per sq. in. decreased as the carbon increased. In general, the annealed specimens showed a higher compression than the cast

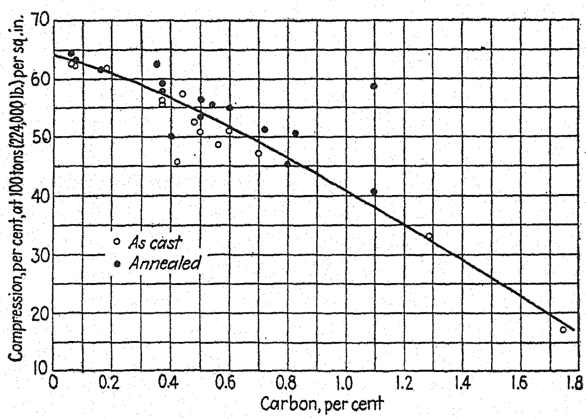

FIG. 9.—Compression of cast carbon steels. (*Arnold.*[4])

specimens of the same carbon content. Cast specimens containing more than 0.70 per cent carbon fractured before the full load was applied, thus supporting the data given in Table 23 in showing that extra-high-carbon steel castings are decidedly low in ductility. This kind of a compression test serves also as an approximate measure of ductility.

E. AUTHOR'S SUMMARY

1. In the present chapter the mechanical properties of three classes of cast steels have been summarized; these classes are: (*a*) low carbon, with less than 0.22 per cent carbon, (*b*) regular grade, with 0.22 to 0.35 per cent carbon, and (*c*) high carbon, with more than 0.35 per cent carbon. For the regular grade it was possible to include data representative of the properties of castings made in commercial production.

2. Depending upon the treatment used, low-carbon castings have a tensile strength varying from about 45,000 to 60,000 lb. per sq. in. for carbon contents under 0.15 per cent, and from about 65,000 to 80,000 lb. per sq. in. for carbon contents of

0.18 to 0.21 per cent. The elongation varies from 30 to 35 per cent for steels of the lower carbon percentages, and from 15 to 30 per cent for steels of 0.18 to 0.21 per cent carbon.

3. In the regular-grade class, tensile strength ranges from 55,000 lb. per sq. in. for cast or annealed material to over 100,000 lb. per sq. in. for quenched and tempered castings. Elongation in general varies inversely with the strength, from 10 to 40 per cent.

4. Typical tensile-strength values of regular-grade cast steels produced commercially are: 75,000 lb. per sq. in. as cast, 75,000 lb. per sq. in. as annealed, and 90,000 lb. per sq. in. as quenched in water and tempered just below the critical temperature. The corresponding average elongations are 20, 30, and 25 per cent. Regular-grade cast steels with these representative properties constitute the principal tonnage of steel castings and by proper heat treatment can be produced to meet practically all commercial specifications. Several investigators have shown that quenching and tempering may be used even with fairly complex castings without danger of distortion.

5. Castings containing 0.36 to 0.50 per cent carbon have a tensile strength of 75,000 to 125,000 lb. per sq. in. and an elongation of 25 to 5 per cent depending upon the heat treatment. The ductility of this grade as cast is low; hence, there is no apparent advantage in using these higher carbon steels unless the best combination of strength and ductility is developed by a suitable heat treatment. The high-carbon castings are used where high strength and wear resistance are desired, as in rolls, dies, or machine tools.

6. A representative tensile strength of 0.40 per cent carbon centrifugally cast steel, annealed and quenched and tempered, is 100,000 to 108,000 lb. per sq. in. with 14 to 16 per cent elongation.

7. Cast steels containing more than 0.60 per cent carbon are extremely brittle. Annealing improves the ductility but in many cases not enough to make this grade of marked commercial importance. Results quoted in this chapter show that the elongation of annealed 0.70 to 1.40 per cent carbon cast steel may be zero and is always below 10 per cent, even after annealing.

8. There are few data on torsional strength of cast steels; those that are available indicate that the ultimate strength in

torsion is lower than in tension. Aside from this, it is not possible to generalize on the relation between tensile and torsion properties of cast steels. Additional data on torsional properties are needed.

9. To judge from the limited data on compressive properties, there seems to be a relation between percentage compression under a load of 100 tons (224,000 lb.) per sq. in. and carbon content, the former decreasing as the latter increases.

10. Further improvement in the properties of steel castings will most likely result from better foundry practice, particularly in avoiding occasional castings of low ductility, and from more extensive use of suitable heat treatments.

CHAPTER IV

MECHANICAL PROPERTIES OF HOT-WORKED CARBON STEELS

Relation between Composition and Strength of Hot-worked Steels—Effect of Variations in Hot-working Practice on Properties—Author's Summary.

About 97 per cent of all the steel made in the United States is cast into ingots which are then processed into various finished or semifinished forms by some method of hot working—mostly by rolling, occasionally by pressing or forging. Of this great tonnage only about 6 or 7 per cent can be classed as alloy steel, the rest being plain carbon steel containing no intentionally added element except carbon and manganese.

In the United States most of the carbon-steel tonnage is produced by the basic open-hearth process as shown by the following percentages for four typical years:

Process	Percentage of production			
	1916	1922	1928	1933
Basic open-hearth	70.1	80.9	85.1	87.2
Acid open-hearth	3.0	1.5	0.9	0.9
Acid Bessemer	26.4	17.0	13.1	10.6
Electric	0.3	0.5	0.9	1.3
Crucible	0.2	0.1	Small fraction	

The increase in the production of basic open-hearth steel, principally at the expense of Bessemer steel, is noteworthy.

42. Hot Working and Cold Working.—Like many of the metals and alloys, iron and iron-carbon alloys are malleable from below room temperature to the melting point, and, following the general rule, the malleability increases with the temperature. An exception to this is commercially pure ingot iron which, as is well known, cannot be rolled or forged in the temperature

range 900 to 1065°C. (1650 to 1950°F.). This is discussed briefly in Chapter XII and also by Epstein (Volume I, page 353, of this monograph). There is no sharp dividing line between hot and cold working; some authorities attempt to place this line at the recrystallization or the critical temperature; most mill men, on the contrary, use the term cold working to denote mechanical treatment at or slightly above room temperature, and hot working for mechanical treatment at all higher temperatures. In hot rolling or forging, a large part of the tonnage is worked at temperatures above the lower critical, the A_1 temperature, and is usually allowed to cool in air from the finishing temperature. There are, however, many hot-working operations in which the finishing temperature is considerably below A_1.

Production percentages of hot-worked steel for three typical years are as follows:

Finished or semifinished product	Percentage of production		
	1922	1928	1933
Rails and track accessories	11.1	9.1	3.7
Plates	12.6	10.5	7.0
Black plate for tinplate	4.9	4.8	11.8
Other sheet and strip	14.2	21.5	30.2
Wire rods	9.7	8.3	12.2
Structural shapes	11.0	11.1	6.7
Bars	17.2	19.0	15.7
Pipe, skelp, etc	10.7	11.9	9.3
Other products	8.6	3.8	3.4

The majority of the above products—rails, plates, structural shapes, and some others—is hot rolled into the finished sections and receives no treatment other than that which naturally results from the rolling;* for example, grain refinement in working and, for the smaller sections, some hardening upon air cooling if the carbon is high enough and if the finishing temperature is above A_1. Other products such as bars may be hot worked further by rolling or forging; the resulting finished section may then be used without further treatment or may be heat treated in various ways.

* A considerable tonnage of rails is now normalized or subjected to controlled cooling from the finishing temperature; or the ends are hardened.

Other semifinished products such as some sheet, and all wire rods, are worked into finished form by cold drawing or cold rolling; the finished product may or may not be heat treated depending upon the use for which the product is designed.

A. RELATION BETWEEN COMPOSITION AND STRENGTH OF HOT-WORKED STEELS

The mechanical properties of steel as hot worked are dependent chiefly upon the composition, especially the percentage of carbon. They are also dependent, but to a less degree, upon the amount of reduction and the direction of working, the finishing temperature and the method of cooling from this temperature, and the size of the section.

In considering the mechanical properties of carbon steels it should be realized that two steels of apparently the same chemical composition (as usually determined by standard methods of analysis) may occasionally show a considerable difference in properties. This has been discussed at some length in Volume I of this monograph (see Chapters I, XI, and XII). As the causes for such differences in properties are known imperfectly, if at all, all that can be done in this and the next two chapters is to indicate what are believed to be, within certain limits, the most probable or representative properties of a steel containing a given amount of carbon after it has been subjected to a certain mechanical or thermal treatment or a combination of the two.

43. Effect of Carbon Content.—There are relatively few data on the effect of carbon content on the tensile properties of hot-worked commercial iron-carbon alloys. Considering that hundreds of thousands of tensile tests have been made by the steel works of this country on steels ranging in composition from ingot iron with 0.02 per cent or less carbon to forging bars and tool steels with 1.00 to 1.50 per cent carbon, it is unfortunate that a statistical analysis of these tests has not yet been made to show more accurately the effect of carbon than can be shown now. The importance of statistical analysis, the application of statistical methods to analyzing mechanical properties in Germany, and the lack of such application in this country have been discussed in Chapter I.

All of the available published reports on the effect of carbon on the mechanical properties of iron, with the exception of the

German statistical data (see page 86), are more than 10 years old. The data of four of these early investigators are summarized in Fig. 10. The compositions of the steels used by Nead[35] and by Langenberg[73,74] are as follows:

Composition, per cent					
C	Mn	P	S	Si	Cr
0.14	0.45	0.018	0.035	0.131	Nil
0.18	0.56	0.024	0.043	0.132	Nil
0.32	0.51	0.009	0.027	0.128	Nil
0.46	0.40	0.020	0.050	0.144	Nil
0.49	0.60	0.013	0.028	0.127	Nil
0.57	0.65	0.012	0.028	0.167	Nil
0.71	0.67	0.027	0.035	0.147	Nil
0.83	0.55	0.018	0.028	0.152	Nil
1.01	0.39	0.016	0.029	0.160	Nil
1.22	0.34	0.025	0.031	0.181	Nil
1.39	0.20	0.015	0.029	0.191	Nil
1.46	0.20	0.011	0.035	0.133	0.35

The process by which these steels were made is not given; from the composition a guess may be hazarded that they were either electric or crucible melted. Tensile specimens were approximately standard: 0.5 in. in diameter and of 2 in. gage length, and were machined from 1-in. hot-rolled rounds. The steels used by Brinell, as reported by Wahlberg,[5] and those described by Roberts-Austen and Gowland in the Sixth Report to the Alloys Research Committee[8] were also commercial alloys and contained the usual impurities or addition elements found in carbon steels. Brinell's steels were made by the acid open-hearth process, and the others, from Sheffield, were probably crucible steels. The tensile specimens used by Brinell had a diameter of 18 mm. (0.71 in.) and a gage length of 100 mm. (3.94 in.); those used in obtaining the results given in the Alloys Report[8] had a diameter of either 7/16 in. or 1/2 in. and a gage length of 2 in. There are no details of the rolling practice or finishing temperatures given by any of these investigators.

There is a surprisingly good agreement between the values obtained by Nead and those obtained by Brinell. Tensile-strength values given in the Alloys Report are in line with

those obtained by Nead and Brinell, but values of reduction of area and elongation are much lower for the low-carbon steels.

As Fig. 10 shows, tensile strength, yield strength, and Brinell hardness increase as the carbon content increases; the curves for

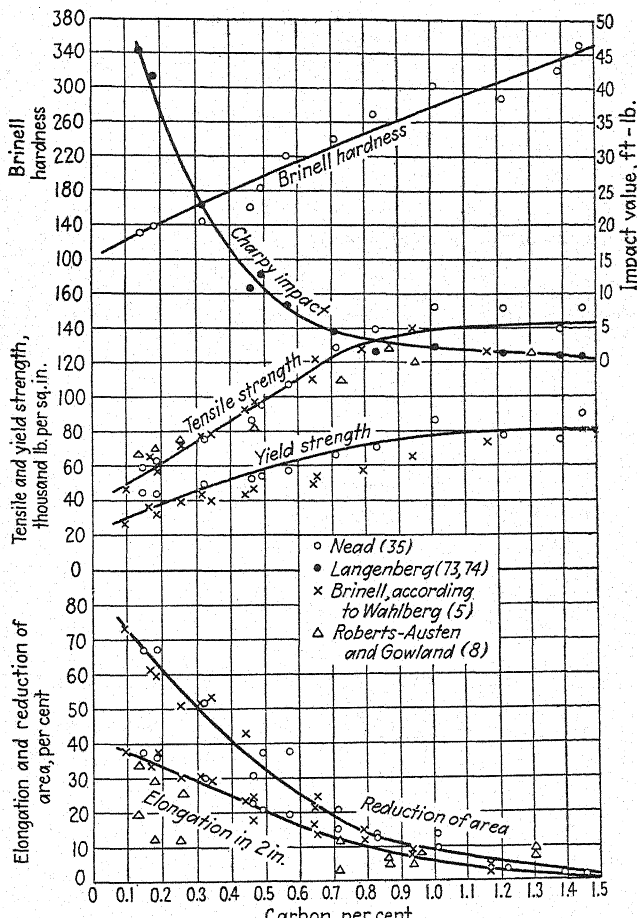

FIG. 10.—Effect of carbon content on mechanical properties of hot-worked steels.

tensile strength and Brinell hardness are almost parallel for carbon percentages of 0.05 to 0.75; likewise tensile- and yield-strength curves are almost parallel for carbon percentages of 1.0 to 1.5. As the carbon increases, the elongation, reduction of area, and Charpy impact decrease sharply.

A statistical analysis of mechanical properties of more than 5000 commercial heats of carbon steel containing between 0.1 and 0.6 per cent carbon has been made by Daeves and others in Germany; a summary of part of the results has been reported in the "Werkstoff-Handbuch Stahl und Eisen."[815] These are given in Table 26, which also gives the corresponding values taken from the curves of Fig. 10. Considering the difference in

TABLE 26.—COMPARISON OF MECHANICAL PROPERTIES OF CARBON STEELS AS CORRELATED IN FIG. 10 WITH RESULTS OF GERMAN STATISTICAL ANALYSES

Carbon, per cent	Tensile strength, lb. per sq. in.		Yield strength, lb. per sq. in.*		Elongation, per cent		Reduction of area, per cent	
	From Fig. 10	German statistical analyses	From Fig. 10	German statistical analyses	From Fig. 10†	German statistical analyses‡	From Fig. 10	German statistical analyses
0.1	49,000	52,000	30,000	38,000	37	29	73	65
0.2	61,000	64,000	38,000	44,500	32.5	24	61	59
0.3	73,000	75,500	46,000	51,000	28	19	50	53
0.4	85,000	87,500	52,000	58,000	24	17	41	44
0.5	97,000	99,500	58,000	64,500	20	15	33	34
0.6	109,000	111,000	62,000	71,000	16	13	26	24

* Values in Fig. 10 mostly reported as yield point, German values as Streckgrenze.
† Mostly on a gage length of 2 in.
‡ On a gage length of 100 mm. (3.94 in.).

gage length, and assuming that Fig. 10 shows with reasonable accuracy the effect of carbon, it may be concluded that carbon steels as made in Germany have approximately the same elongation and reduction of area and the same, or slightly higher, tensile strength as the steels made in this country and in England. The yield strength of the German steels, however, is higher.* Data on the effect of carbon contents between 0.02 and 0.30 per cent on the tensile and yield strengths of hot-rolled strip of five different gages are given on page 106.

* Assuming that the results of the German statistical analyses represent true average mechanical properties of commercial carbon steels, it may, of course, be concluded with equal validity that the yield strengths of the steels whose properties are plotted in Fig. 10 were incorrectly determined or at least do not correspond approximately to the "Streckgrenze."

44. Formulas for Calculating Tensile Strength of Carbon Steels.—In the period between 1892 and 1922 much time and effort were spent in attempting to work out a formula for calculating tensile strength from composition. The work of Campbell[10,12] and Webster,[95,108] in the United States, and of

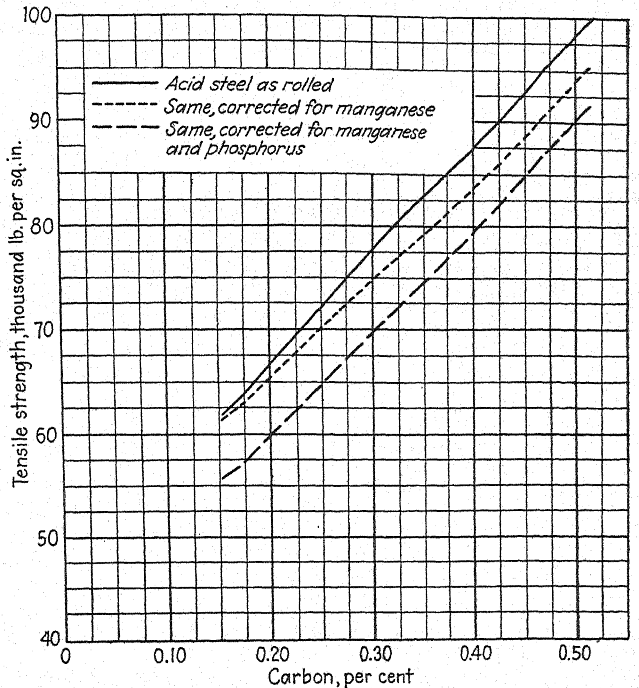

FIG. 11.—Effect of carbon content on the tensile strength of acid steel. (*Campbell.*[10])

McWilliam,[49] in England, is well known and should have brief mention.

Campbell determined tensile properties of a large number of acid and basic open-hearth steels of varying carbon content. The tensile strengths are plotted as solid lines in Figs. 11 and 12. In order to determine the effect of carbon it was necessary to correct for manganese and phosphorus. In doing this Campbell arrived at a correction of 1000 lb. per sq. in. for each 0.01 per cent phosphorus, and a variable unit correction for manganese which for acid steel ranged from 160 to 9600 lb. per sq. in., and

for basic steel from 500 to 7500 lb. per sq. in. depending upon the carbon and manganese percentages.

In Figs. 11 and 12 the dotted and dashed lines show the tensile strengths versus carbon percentages after these corrections. Each 0.01 per cent carbon was found to increase the tensile

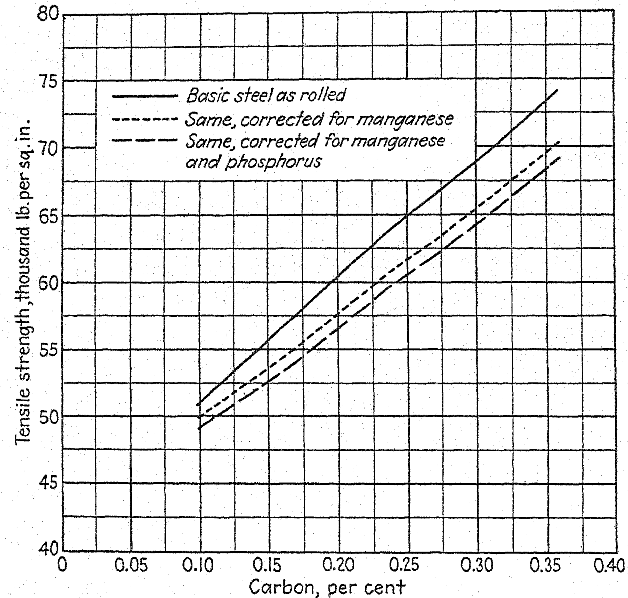

FIG. 12.—Effect of carbon content on the tensile strength of basic steel. (*Campbell.*[10])

strength 1000 lb. per sq. in. for acid steel, and 770 lb. per sq. in. for basic steel. The formulas thus became:
For acid steel:

$$\text{Tensile strength} = 40{,}000 + 1000\text{C} + 1000\text{P} + x\text{Mn}$$

For basic steel:

$$\text{Tensile strength} = 41{,}500 + 770\text{C} + 1000\text{P} + y\text{Mn}$$

in which the values 40,000 and 41,500 were taken as the tensile strength of carbon-free acid and basic irons. The carbon, phosphorus, and manganese are in units of 0.01 per cent; x and y are variables depending upon the carbon and manganese percentages, as noted above, and are given by Campbell in tables.

Another formula was worked out by Webster[95,108] from the data plotted in Figs. 13 and 14. The gap between the curves for the 0.55 and 0.60 per cent carbon steels (Fig. 13) was ascribed to the presence of more silicon in the higher carbon material. The value for the increase in tensile strength for each 0.01 per

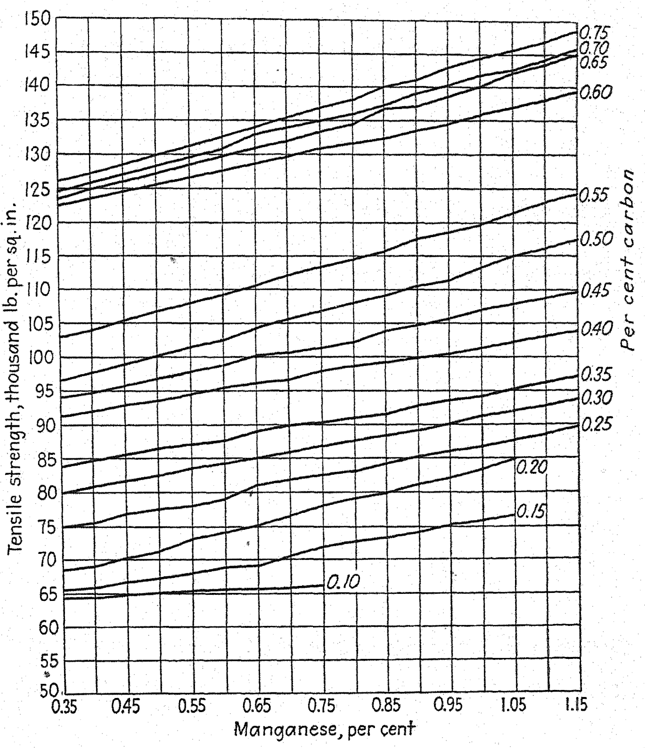

FIG. 13.—Effect of carbon and manganese on the tensile strength of acid Bessemer steel. (*Webster.*[95])

cent carbon was found to be 800 lb. per sq. in. Webster accepted 38,000 lb. per sq. in. as the tensile strength of carbon-free iron and gave the formula

$$\text{Tensile strength} = 38{,}000 + 800\text{C} + x\text{P} + y\text{Mn}$$

in which x varies from 800 to 1500 lb. per sq. in. depending upon the carbon. The correction for manganese (y) also varies according to the carbon.

In England, McWilliam [49] used the formula

$$\text{Tensile strength} = 38{,}000 + 800C + 100Mn + 1000P$$

but found it valid only in a carbon range of 0.20 to 0.26 per cent.

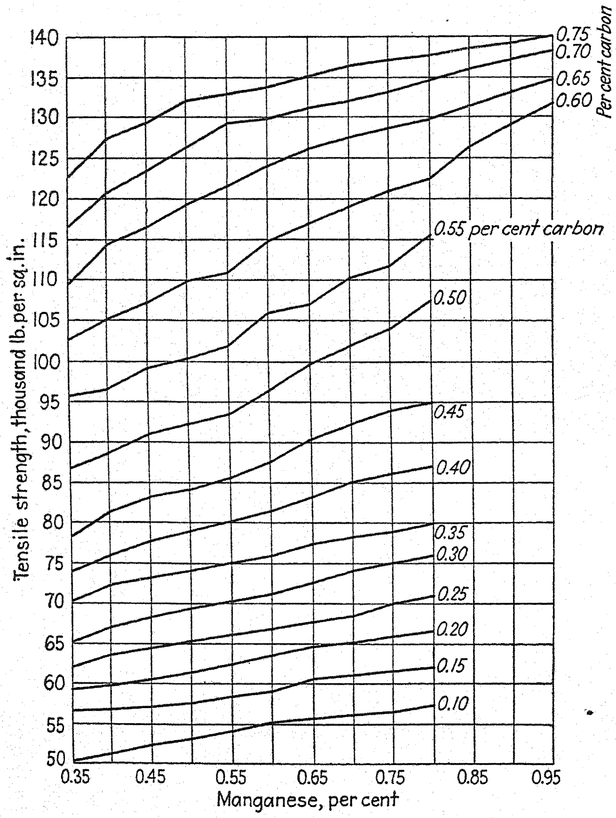

FIG. 14.—Effect of carbon and manganese on the tensile strength of basic open-hearth steel. (*Webster*.[95])

He then worked out a more complex formula

$$\text{Tensile strength} = 38{,}000 + C[800 + 4(C - 20)] + Mn[100 + 2(C - 20)] + 1000P + 120Si$$

which, according to Nead,* gives satisfactory results on steels of about 1-in. section if the manganese is not more than 1 per cent. Nead recommended that to the value obtained by McWilliam's

* Private communication.

formula 1000 lb. per sq. in. be added for each ⅛ in. on sections lighter than ¾ in.

In the recent publication of the Bethlehem Steel Company,[791] three formulas for calculating the tensile strength of carbon steels are given. These are:

$$\text{Tensile strength} = 38{,}800 + 650\text{C} + 90\text{Mn} + 4\text{MnC} + 1000\text{P}$$
$$\text{Tensile strength} = 37{,}430 + 950\text{C} + 85\text{Mn} + 1050\text{P}$$
$$\text{Tensile strength} = 38{,}800 + (650 + 4\text{Mn})\text{C} + 90\text{Mn} + 1000\text{P}$$

In all of these formulas, the values for carbon, manganese, and phosphorus to be inserted are the percentages in which the decimal point has been moved two places to the right (thus 0.30 per cent carbon becomes 30C and 0.04 per cent phosphorus becomes 4P).

Campbell as early as 1905 showed conclusively that, while he could calculate accurately the tensile strength of most rolled carbon steels (of less than 0.50 per cent carbon) from the composition, instances would crop up where the discrepancy between the calculated and actual strengths was as much as 6000 lb. per sq. in., and errors of 1000 to 2000 lb. per sq. in. were fairly common. Despite this, the attempts to work out an accurate correlation between composition and strength persisted until 1923.

Notwithstanding the apparent impossibility of devising an absolute relationship which will hold for all low- and medium-carbon steels, empirical formulas such as the McWilliam or Bethlehem Steel Company formulas given above are still used successfully in many tonnage mills and are entirely satisfactory where years of standardized mill practice have resulted in at least partly eliminating the effect of such factors as widely varying manganese and phosphorus contents, finishing temperatures of rolling, cleanliness, and others.

If all other conditions are equal, there is probably a fairly straight-line relationship between carbon, from 0.05 to 0.60 per cent, and tensile strength and hardness, in which each 0.01 per cent carbon increases the tensile strength about 1000 lb. per sq. in. and the Brinell number by 2.

45. Variation of Ductility with Strength.—In order to determine the relationship between ductility and tensile strength of commercial carbon steels Daeves[361] made a statistical study of

data obtained from the laboratory reports of several German steel mills and data published in Sweden. The test results used were for steels rolled to various sections, and should be typical of the materials produced by large-scale methods. For each value of tensile strength two frequency curves were plotted, one

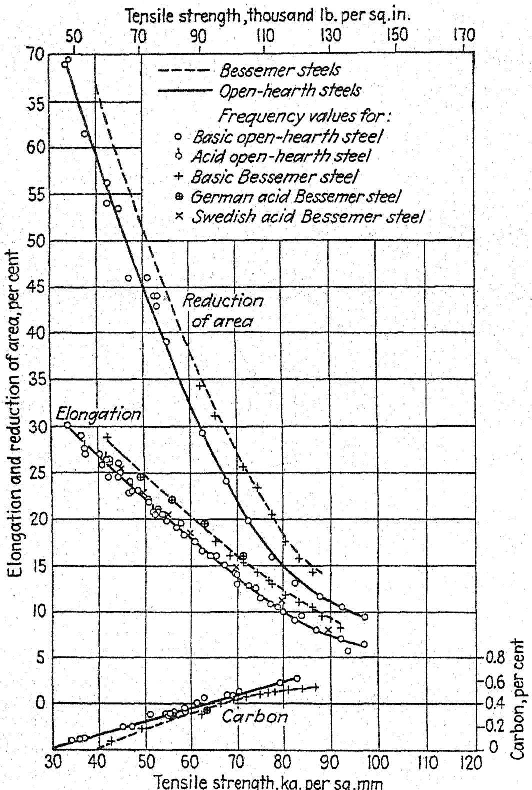

FIG. 15.—Most-probable tensile properties of hot-rolled carbon steels made by various processes. (*Daeves.*[361])

for elongation—in 100 mm. (3.94 in.)—and the other for reduction of area. Most-probable values of elongation and reduction of area as obtained from these frequency curves are shown in Fig. 15. Each point plotted in Fig. 15 represents from 50,000 to over 100,000 values. The most probable carbon content for a given tensile strength is also shown in the figure.

According to Daeves' statistical analyses, steel made by the basic Bessemer process has a slightly higher ductility for a given

tensile strength than steel made by the basic open-hearth process. The few values for steels made by the acid open-hearth process fall on the tensile-strength versus elongation curve for basic open-hearth steels.

One outstanding difference between Bessemer steels and basic open-hearth steels is the phosphorus content, the Bessemer steels usually containing more phosphorus. Bessemer steels also contain more nitrogen than open-hearth steels, which, like phosphorus, affects the properties. Data plotted in Fig. 16

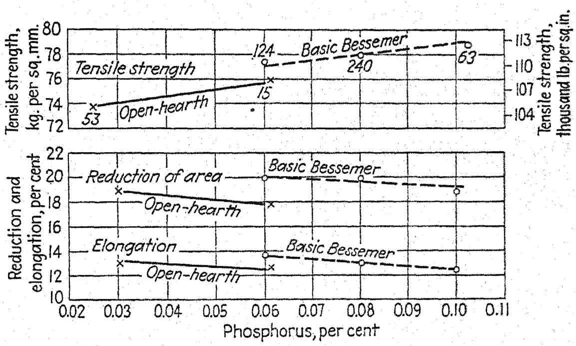

FIG. 16.—Tensile properties of hot-rolled basic Bessemer and basic open-hearth steels containing different amounts of phosphorus. The numerals at the plotting points indicate the number of determinations for the frequency curves. (*Daeves.*[361])

indicate that for the same phosphorus content basic Bessemer steels have higher values of tensile strength, elongation, and reduction of area than basic open-hearth steels.

Figure 17, also from Daeves, shows maximum and minimum as well as most probable values of elongation for basic open-hearth steels. The difference between maximum and minimum values may seem large unless it is realized that values were obtained from materials of various cross-sections and produced at different mills.

B. EFFECT OF VARIATIONS IN HOT-WORKING PRACTICE ON PROPERTIES

In general, the mechanical properties of carbon steel are improved by hot working. This improvement may be wholly due to the specific effects of plastic deformation, or it may be due to a combination of deformation and coincident heat treat-

ment which the material may undergo if the finishing temperature is above the critical, if the carbon is high enough, and if the section is relatively small.

46. Influence of the Direction of Hot Working on Properties.—One of the chief advantages of hot working is that it breaks up the coarse cast structure. In addition it causes fibering by elongating the dendrites in the direction of working. Inclusions

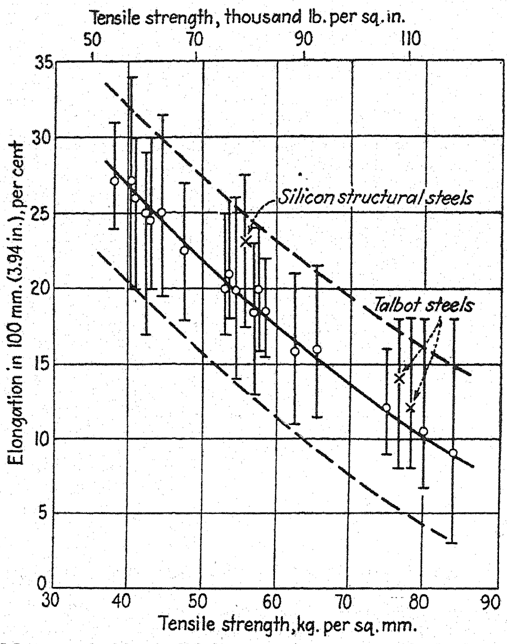

FIG. 17.—Most probable (solid line) and extreme (dashed lines) values for elongation of hot-rolled basic open-hearth steels of different tensile strength. (*Daeves.*(361))

which are plastic at rolling and forging temperatures—many of them are—are also elongated into fibers. The elongation of grains and inclusions tends to improve the mechanical properties when the material is subjected to stresses acting longitudinally (in the direction of working) and also results in inferior properties, especially in values for elongation, reduction of area, and impact resistance, in specimens cut transversely ("across the grain," at right angles to the direction of working).

Even in carbon steels melted and refined by the best present-day practice there is usually a slight but appreciable difference

in ductility between longitudinal and transverse specimens. This difference is so accentuated by inclusions which are elongated in rolling that it is not uncommon for the impact value of

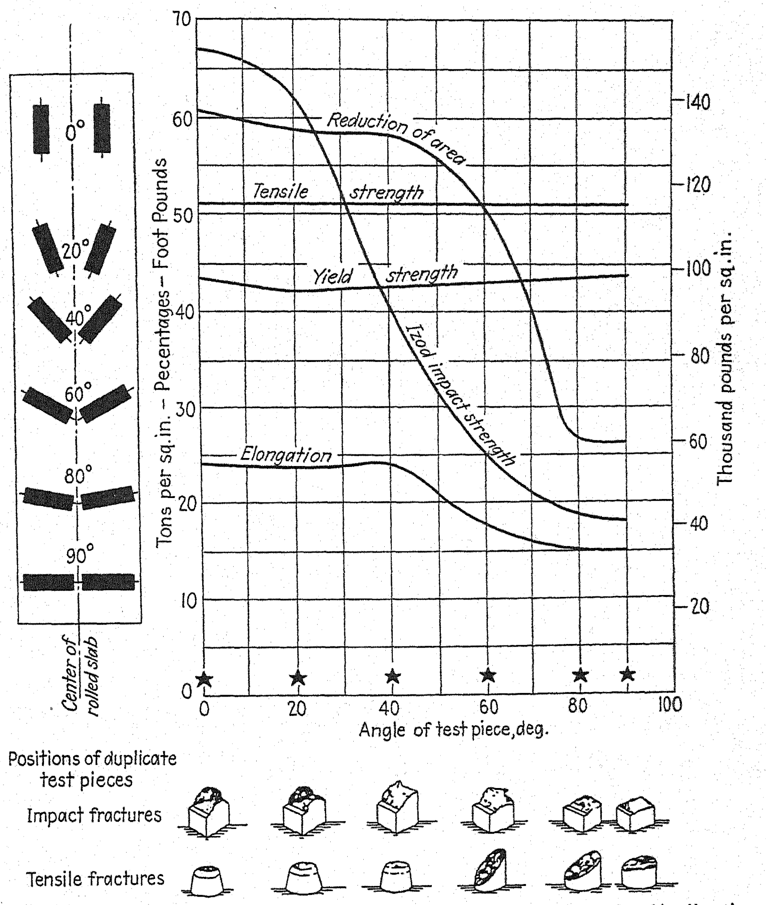

FIG. 18.—Relation between the mechanical properties and the angle of inclination of the fibers to the axis of the test piece. (*Aitchison*[64] *after Brearley.*)

a relatively dirty steel to be 40 ft-lb. in a longitudinal specimen and less than 5 ft-lb. in a specimen cut transversely to the direction of working. Sheet and plate, which are cross-rolled, do not show such great directional differences in properties as bars, rails, structural shapes, or other sections which are elongated in one direction only.

The effect of inclusions, in intensifying differences in longitudinal and transverse mechanical properties, has been discussed briefly in Volume I of this monograph (see page 362) from which Fig. 18 has been reproduced. Brearley's tests, quoted by Aitchison,[64] were made on an alloy steel, but the behavior is general, as will be shown by further examples on carbon steel given below. As shown in Fig. 18, tensile and yield strengths are unaffected by the angle of inclination of the fiber, while the properties representing ductility—elongation, reduction of area, and impact resistance—are sharply reduced to a minimum at right angles to the direction of rolling.

Charpy[46] studied the effect of hot working on a basic open-hearth steel, presumably of forging quality, by tests on specimens taken longitudinally and transversely. The ingot used was 355 mm. (14 in.) square; one portion was forged until the reduction was 59.8 per cent, and another until reduction was 87.6 per cent. Specimens from the original ingot and from the forged sections were quenched in water from 850°C. (1560°F.) and tempered at 650°C. (1200°F.) apparently so that all tests would be made on material of the same structural condition so far as possible. The results obtained are as follows:

Reduction in hot working, per cent	Tensile strength, lb. per sq. in.	Elongation, per cent*	Reduction of area, per cent	Charpy impact value, m-kg.
		Longitudinal		
0	101,500	16	33	3.3
59.8	99,700	18	33	3.5
87.6	103,500	23	60.7	9.1
		Transverse		
0	102,000	14	22	3.4
59.8	100,800	11	27	2.0
87.6	98,800	4	8	1.5

* Gage length not given.

As Stead and Brearley brought out in discussing Charpy's paper, mere fibering by elongation of the dendrites would not produce such poor transverse properties, so that the steel must

have contained a large number of slag inclusions. If, as claimed by Brearley, the ratio of the impact values for longitudinal and transverse tests is a measure of the comparative freedom from slag, Charpy's steel must have been rather dirty.

Data on transverse and longitudinal properties which are probably more representative of the directional differences in commercial carbon steel were reported by Roš.[406] The tests were cut from the flange and web sections of large I-beams, rolled from basic Bessemer steel of the following composition:

Steel	Composition, per cent				
	C	Mn	P	S	Si
A	0.05	0.64	0.05	0.05	0.01
B	0.07	0.64	0.045	0.04	0.00
C	0.15	1.03	0.06	0.05	0.01

Results are given in Table 27. The elongation and reduction of area of the transverse specimens are appreciably lower than the same values of the longitudinal tests but not nearly so low as Charpy's values (see page 96). The impact resistance of the specimens having a notch parallel to the rolling direction (transverse) is less than half that of the longitudinal specimens. The strength of the flange in both directions is less than that of the web. This may be attributed to two factors. First, in rolling I-beams, the web contains the metal from the center and most strongly segregated part of the ingot, hence strength would be higher and ductility lower than those of the less segregated flange. Second, and probably not so important, the flange is thicker than the web.

The difference in ductility in longitudinal and transverse specimens is also apparent in plates which are rolled directly from the slab. This has been concisely summarized by Camp and Francis[150] who stated:

> If a slab be rolled in one direction only and a longitudinal and transverse test piece be cut from the resulting plate, little difference in tensile strength will be observed in pulling the two tests but a marked difference in ductility will be found. Thus the longitudinal piece will give 4 to 7 per cent greater elongation than the transverse piece, and 10 to 15 per

cent greater reduction in area. This difference is to be attributed to the fiber structure produced by the one-way rolling, for it is not noticeable in cross-rolled plates.

Differences in the ductility between longitudinal and transverse specimens are reduced to a minimum and sometimes completely eliminated in the manufacture of carbon- and alloy-steel forgings for ordnance guns, shells, and the like. As stresses in such sections are chiefly transverse, specifications require that superior

TABLE 27.—LONGITUDINAL AND TRANSVERSE PROPERTIES OF HEAVY I-BEAMS*

Steel	Carbon, per cent	Location	Direction†	Tensile strength, lb. per sq. in.	Yield strength, lb. per sq. in.	Elongation, $l = 5d$, per cent	Elongation, $l = 10d$, per cent	Reduction of area, per cent	Charpy impact, m-kg. per sq. cm.	Brinell hardness
A	0.05	Flange	L	54,500	33,100	36.0	28.0	72	17.2	108
		Flange	T	57,200	35,800	27.6	22.7	47	7.5	
		Web	L	59,900	44,400	26.8	20.6	59	15.6	122
		Web	T	63,900	49,200	26.0	19.7	36	5.8	
B	0.07	Flange	L	53,800	37,600	34.0	25.0	72	18.0	113
		Flange	T	54,600	33,600	31.0	25.2	55	7.2	
		Web	L	58,300	44,200	26.6	18.9	67	17.4	132
		Web	T	61,300	51,300	22.4	18.1	33	4.6	
C	0.15	Flange	L	74,500	43,000	36.7	27.3	73	18.1	137
		Flange	T	72,400	38,100	16.3	12.5	21	6.5	
		Web	L	75,700	53,900	29.7	22.5	57	14.4	154
		Web	T	77,800	55,900	20.0	15.7	35	4.0	

* Roš.[406]
† L = longitudinal; T = transverse.

transverse properties shall be attained. This is accomplished by very clean steel, careful forging, and a complex heat treatment which tends to destroy completely any vestige of fiber.

47. Effect of Hot-working Temperature on Properties.—Freeman and Derry[117] studied the influence of the following five factors on the mechanical properties of hot-rolled plate; initial temperature of rolling, finishing temperature, total reduction, reduction per pass, and rolling speed. The work was done with a steel containing 0.5 per cent carbon, on the 16-in. mill at the National Bureau of Standards. The results indicated that

total reduction and finishing temperature had the greatest influence on the properties of the hot-rolled plate. Normalizing tended to obliterate any differences resulting from variations in rolling conditions.

A study reported by Oberhoffer, Lauber, and Hammel[40] is of interest in showing the effect of hot-working temperature and variations in cooling after forging on the properties of low-, medium-, and high-carbon steels. The compositions of the steels used were:

Steel	Composition, per cent				
	C	Mn	Si	P	S
A	0.10	0.40	0.02	0.012	0.024
B	0.40	0.70	0.09	0.014	0.031
C	0.77	1.28	0.21	0.030	0.029

Blocks 50 × 50 × 140 mm. (1.97 × 1.97 × 5.5 in.) were cut from sound ingots and forged to sections 25 × 25 mm. (0.98 × 0.98 in.) in cross-section, which represents a reduction in cross-sectional area of 75 per cent. Mechanical properties of the bars are shown in Fig. 19, which gives the temperatures at the beginning and end of the forging operation. The elongation was measured over a gage length of 100 mm. (3.94 in.). Impact tests were made on 10 × 30 × 160-mm. (0.39 × 1.18 × 6.3-in.) specimens. Curves labeled 1 are for samples cooled slowly after forging and those labeled 2 are for samples cooled in air. Curves labeled 3 indicate the properties of samples annealed after forging; the annealing temperatures were 900°C. (1650°F.) for steel *A*, 850°C. (1560°F.) for steel *B*, and 700°C. (1290°F.) for steel *C*.

According to the data shown in Fig. 19, the properties of the 0.77 per cent carbon steel were affected very little by varying the hot-working temperature. Decreasing the forging temperature increased the yield strength and impact resistance but not the tensile strength or hardness of the 0.40 per cent carbon steel. Elongation and reduction of area were but slightly affected. The tensile and yield strengths, elongation, and Brinell hardness of samples *A*1 and *A*2 of the low-carbon steel were affected

by changes in forging temperature; on the contrary the annealed specimen $A3$ was unaffected. The lack of uniformity in the curves shown in Fig. 19 makes it impossible to generalize upon

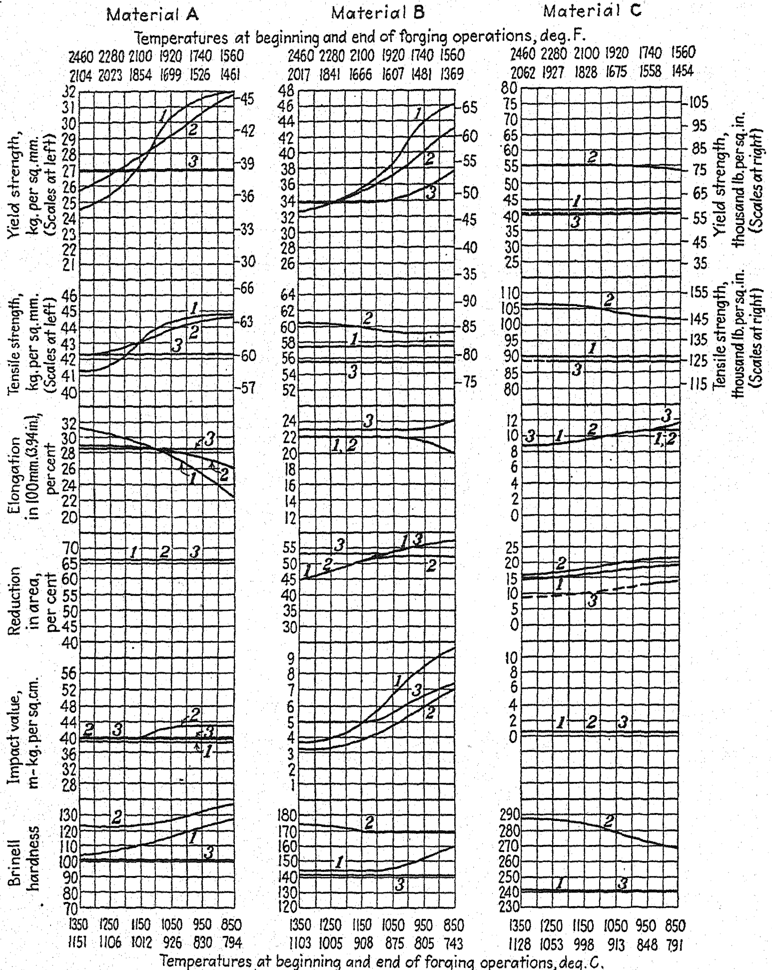

FIG. 19.—Effect of forging temperatures on the mechanical properties of low-, medium-, and high-carbon steels. (*Oberhoffer, Lauber, and Hammel.*[40])

the effect of hot-working temperature on carbon steel of varying carbon content.

In some tests undertaken to determine the influence of rolling temperature on the properties of rails, Stumper[346] rolled basic

Bessemer rails containing 0.40 per cent carbon and 1 per cent manganese. Finishing temperatures of from 950 to 1150°C. (1740 to 2100°F.) were used. As the finishing temperature was raised, tensile strength and elongation decreased, but Brinell hardness was almost independent of rolling temperature, averages for the whole rail section ranging from 211 to 217. Notched-bar impact resistance at 25°C. (75°F.) increased as lower finishing temperatures were used. Average properties of the rails are given in Table 28.

TABLE 28.—PROPERTIES OF RAILS (0.40 PER CENT CARBON) ROLLED AT DIFFERENT TEMPERATURES*

Finishing temperature		Tensile strength, lb. per sq. in.	Yield strength, lb. per sq. in.	Elongation in 200 mm. (8 in.), per cent	Charpy impact value,† m-kg. per sq. cm.	
°C.	°F.				(*a*)	(*b*)
			Head of rail			
950	1740	103,800	59,500	15.0	0.73	1.29
1050	1920	103,800	56,800	14.8	0.85	1.14
1150	2100	101,000	55,300	14.6	0.74	0.78
			Web of rail			
950	1740	102,300	62,600	15.3		
1050	1920	101,100	58,700	14.9	0.59	
1150	2100	99,100	59,000	15.2	0.46	
			Base of rail			
950	1740	102,700	64,700	17.5	0.73	1.83
1050	1920	106,200	67,000	16.0	0.81	1.03
1150	2100	99,300	60,600	16.5	0.68	0.95

* Stumper.(346)

† 10 × 10 × 60-mm. (0.4 × 0.4 × 2.4-in.) samples with notch 5 mm. (0.2 in.) deep. (*a*) Tests at 0°C. (32°F.); (*b*) tests at 25°C. (75°F.).

48. Influence of Hot-working Temperatures on Properties of Low-carbon Strip.—A comprehensive investigation on the effect of mill practice and annealing temperatures, to provide data on processing continuous-mill sheets, has recently been completed by

Nead, Mahlie, and Dittrich.* Although most of these data are on the effect of cold rolling and subsequent annealing and are, therefore, discussed in the next chapter, there are a few mechanical-property values which show the effect of finishing temperatures on the rolled and on the rolled and annealed or normalized strips.

The steel was a low-carbon basic open-hearth material, Inland Steel Company specification No. 5, for deep-drawing sheet. The average analysis, and the minimum and maximum, of all the strip tested was as follows:

Element	Percentage		
	Average	Minimum	Maximum
Carbon	0.045	0.03	0.06
Manganese	0.37	0.36	0.39
Sulphur	0.023	0.018	0.028
Phosphorus	0.008	0.007	0.010

The material tested was in the form of hot-rolled 0.080-in. bands. Sheets, 50 in. long, were cut from the various bands, tested for uniformity of gage, pickled, and box annealed. The annealing temperatures reported were actual sheet temperatures. The annealing charge consisted of 60 sheets and was heated for 6 hr. to the annealing temperature, held 4 hr. at temperature, and cooled in 8 hr. Tensile, Olsen ductility, and metallographic specimens were taken longitudinally at the edge and center of the sheet and transversely at the center. Mechanical tests included tensile strength, yield strength (reported as yield point), yield ratio, elongation in 2 in., and Rockwell *B* hardness. Finishing temperatures were 730, 790, and 880°C. (1345, 1450, and 1615°F.) for flat bands and 790°C. (1450°F.) for coiled bands. Results of tensile tests taken longitudinally and transversely at the center are given in Table 29.

Postponing the principal conclusions of Nead and his associates concerning the effect of hot-working temperatures on the proper-

* The complete results of this unpublished investigation were made available for inclusion in this monograph through the courtesy of Wilfred Sykes, Vice-President of the Inland Steel Company.

TABLE 29.—EFFECT OF FINISHING TEMPERATURES ON THE TENSILE PROPERTIES OF 0.08-IN. LOW-CARBON HOT-ROLLED STRIP*

Annealing temperature		Finished at 730°C. (1345°F.), flat			Finished at 790°C. (1450°F.), flat			Finished at 790°C. (1450°F.), coiled			Finished at 880°C. (1615°F.), flat		
°C.	°F.	Tensile strength, lb. per sq. in.	Yield strength, lb. per sq. in.	Elongation in 2 in., per cent	Tensile strength, lb. per sq. in.	Yield strength, lb. per sq. in.	Elongation in 2 in., per cent	Tensile strength, lb. per sq. in.	Yield strength, lb. per sq. in.	Elongation in 2 in., per cent	Tensile strength, lb. per sq. in.	Yield strength, lb. per sq. in.	Elongation in 2 in., per cent
Longitudinal													
Not annealed		56,700	46,700	23.5	49,400	37,800	31.5	46,500	31,600	32.5	55,300	40,600	25.0
595	1100	43,600	27,100	31.0	43,500	27,400	35.0	47,900	34,800	30.5	49,900	33,500	37.0
650	1200	42,300	24,300	32.0	42,000	23,400	35.0	47,600	31,500	26.0	49,000	32,200	34.0
705	1300	41,100	24,100	32.5	39,800	22,100	35.0	40,400	23,400	33.5	47,500	28,000	35.0
760	1400	40,400	25,100	34.3	39,700	22,600	35.0	42,100	26,800	35.0	45,300	31,800	27.5
980†	1800†	47,700	28,200	38.5	46,400	27,600	38.0	48,500	32,890	34.0	50,500	32,200	37.0
980‡	1800‡	45,400	27,500	34.0	46,400	26,500	35.5	47,100	28,300	35.0	49,200	31,400	33.5
Transverse													
Not annealed		59,100	45,900	19.0	52,100	38,200	35.5	50,100	36,000	34.0	54,600	38,700	26.5
595	1100	48,400	35,400	33.5	46,100	27,200	32.5	49,900	33,800	36.0	50,200	29,000	33.5
650	1200	46,200	31,800	31.5	45,600	......	36.0	48,600	32,900	35.0	49,100	30,000	29.0
705	1300	45,500	26,300	30.0	43,900	25,400	34.0	45,300	27,900	30.0	47,600	29,100	35.0
760	1400	44,500	27,000	32.0	44,400	26,100	36.0	44,700	27,200	36.0	48,200	32,300	32.5
980†	1800†	48,600	28,900	35.0	47,900	30,800	36.0	49,600	34,300	37.0	50,700	32,400	37.5
980‡	1800‡	54,400	31,900	34.0	57,800	32,200	38.0	59,100	35,900	33.5	61,000	35,600	38.0

* Nead and associates.
† Normalized only.
‡ Normalized and box annealed at 650°C. (1200°F.)

ties of the cold-rolled and annealed sheet, the following may be concluded from the data given in Table 29:

1. Rolling strip with a finishing temperature of 730°C. (1345°F.) produces grain distortion with higher tensile and yield strengths and a lower elongation than result from a higher finishing temperature. Even after annealing, the elongation is generally lower.

2. Strip flat rolled with a finishing temperature of 790°C. (1450°F.) has slightly lower tensile and yield strengths and markedly higher elongation values than strip flat rolled with a finishing temperature of 880°C. (1615°F.). After annealing or normalizing, tensile and yield strengths of the 790°C. (1450°F.) strip are still slightly lower than those of the 880°C. (1615°F.) strip, but the elongation values are about the same.

3. In general, coiling decreases the tensile and yield strengths of unannealed strip. The yield ratio is markedly decreased.

4. In most cases, the tensile and yield strengths of the transverse specimens are slightly higher than the same properties of the corresponding longitudinal specimens. This is not so evident with the specimens rolled with a finishing temperature of 880°C. (1615°F.) as with those finished colder.

5. For all finishing temperatures the combination of highest strength and highest ductility results from subsequent normalizing or normalizing followed by annealing.

6. As a general average of all tests made on the hot-rolled (not annealed) specimens, Nead and his associates gave the following values for tensile strength and yield ratio:

Finishing temperature		Condition	Tensile strength, lb. per sq. in.	Yield ratio, per cent
°C.	°F.			
730	1345	Flat	55,200	80.4
790	1450	Flat	50,000	76.6
790	1450	Coiled	48,200	70.8
880	1615	Flat	54,000	72.4

49. Effect of the Amount of Reduction on Tensile Properties.—If other things are equal, the greater the reduction in hot working, the higher are the tensile and yield strengths. The effect of the amount of reduction on the tensile strength of hot-rolled commercial Bessemer and basic open-hearth rods is shown on page 105.* Two steels, in the form of 4-in. sq. billets, containing approximately 0.08 per cent carbon and 0.35 to 0.45 per cent manganese were rolled on a continuous mill to the diameters shown.

* Sisco, unpublished data.

The tensile-strength values of the hot-rolled rods (all of which were finished above the critical temperature) were as follows:

Basic open-hearth steel		Bessemer steel	
Diameter of rod, in.	Tensile strength, lb. per sq. in.	Diameter of rod, in.	Tensile strength, lb. per sq. in.
0.594	48,300	0.938	59,000
0.469	50,400	0.689	60,500
0.406	52,000	0.500	63,100
0.312	53,800	0.313	66,000
0.228	57,800	0.250	69,400
		0.207	71,900

Nead* supplied the data, obtained in a large number of tests, plotted in Fig. 20, which shows the tensile and yield strengths of hot-rolled strip of five gages and carbon contents varying from 0.02 to 0.30 per cent. The manganese range of all of the steels was 0.35 to 0.55 per cent. Finishing temperatures for the various gages were as follows:

Gage		Finishing temperature	
Number	in.	°C.	°F.
3	0.250	905	1665
5	0.219	900	1655
9	0.156	855	1570
12	0.109	850	1560
15	0.070	800	1470

The effect of the amount of reduction on the tensile and yield strengths varies (for the specific finishing temperatures given above) with the carbon content; the increase in both of these values is greater as the carbon content increases.

The most that can be said for the data presented on the preceding pages is that the reproducibility of the values reported by the various investigators on the effect of hot-working practice on properties depends almost wholly upon reproducing exactly

* Private communication.

every variable. None of the quoted investigators apparently considered the location of the transformation temperature in

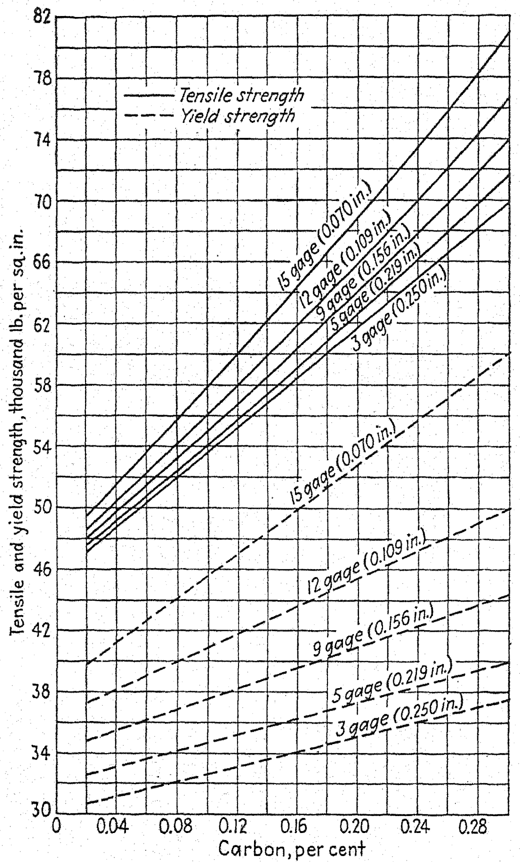

FIG. 20.—Effect of carbon content and amount of reduction in hot rolling on the tensile and yield strengths of strip. (*Nead.*)

his materials. That this is important is apparent from the following comment by John Johnston:*

It is certain that the properties of the finished material depend to a considerable extent on the temperature at which the transformation [gamma to alpha] actually took place; consequently the influence of finishing temperature, in itself, could be discovered only by ensuring that the transformation should go in all cases at the same temperature.

* Private communication.

Otherwise with difference in finishing temperature there is sure to be some difference in transformation temperature on a given cooling bed, and also some difference with different cooling beds and with different sizes of finished material; consequently there will be differences due directly to differences in the actual transformation temperature of the steel. Until this is taken into account, observed differences with different rolling or finishing temperatures are of little real significance.

C. AUTHOR'S SUMMARY

1. There are relatively few data on the mechanical properties of hot-worked steels of varying carbon content. Statistical analyses of mechanical properties have been made in Germany but not in the United States. The application of this analytical method to the thousands of results on file in the works' laboratories of this country is urgently needed.

2. From results which are available for carbon steels of commercial composition, containing 0.10 to 0.60 per cent carbon, it may be concluded that 0.01 per cent carbon increases the tensile strength 800 to 1200 lb. per sq. in., the yield strength 600 to 800 lb. per sq. in., and the Brinell hardness about two units. The exact increase in tensile and yield strengths caused by carbon depends upon a number of factors including the total percentage of carbon, manganese, and phosphorus, and upon the melting and forging or rolling practice.

3. Formulas have been worked out for calculating the tensile strength of hot-worked carbon steels from the composition. Although these are not dependable if absolute accuracy is required, they have considerable value in mills where practice is standardized and where, consequently, some of the variations in strength owing to factors other than composition are not present, or are of little influence.

4. According to German statistical analysis, basic Bessemer steels have slightly higher ductility for a given tensile strength than basic open-hearth material. Acid-process steels apparently differ but little in static tensile properties from basic open-hearth steels. These results need further confirmation.

5. The direction of hot working affects the tensile properties. Specimens cut longitudinally to the direction of hot working will, in general, show a higher elongation, reduction of area, and impact resistance than transverse specimens. Tensile strength

is about the same. The difference in directional properties, present apparently even in very clean hot-worked steels, is accentuated by inclusions which are plastic at hot-working temperatures and are thus elongated into fibers by the mechanical work.

6. There are too few data available to permit broad generalizations on the effect of beginning or finishing temperatures in hot working on the tensile properties. The results of one investigation indicate that, under the specific experimental conditions used, decreasing the forging temperature of a 0.40 per cent carbon steel increases the yield strength and impact resistance but has little or no effect on the tensile strength and hardness; it has only a slight effect on elongation and reduction of area. Varying the forging temperature of a 0.77 per cent carbon steel has, according to the reported results, an almost negligible effect on all the mechanical properties. The results of this same investigation on low-carbon steel showed no uniform trend. The data reported by another investigator indicate that in 0.40 per cent carbon basic Bessemer rails increasing the finishing temperature decreases the tensile strength and elongation slightly. The impact resistance at 25°C. (75°F.) is also decreased.

7. According to the results of an unpublished investigation by Nead and associates, rolling low-carbon strip steel (0.05 per cent carbon) with a finishing temperature of 730°C. (1345°F.) produces grain distortion with a resulting higher tensile strength, a higher yield strength, and a lower elongation than when finishing at higher temperatures. This lower elongation persists after annealing. Flat-rolled strip, finished at 790°C. (1450°F.), has slightly lower tensile and yield strengths and a markedly higher elongation than strip finished at 880°C. (1615°F.). Coiling strip rolled with a finishing temperature of 790°C. (1450°F.) decreases the tensile and yield strengths slightly and the yield ratio markedly. Strip rolled with a finishing temperature of 790°C. (1450°F.) or lower has slightly higher tensile and yield strengths in transverse specimens than in longitudinal specimens. Elongation values are mostly unaffected by the direction of the specimen.

8. If other things are equal, the greater the reduction in hot working, the higher are the tensile and yield strengths. Data are

presented which indicate that, for low-carbon strip, finished at certain stated temperatures, the increase in tensile and yield strengths varies with the carbon, increasing as the percentage of carbon increases.

9. The foregoing conclusions on the effect of variations in hot-working practice on mechanical properties should be considered as valid only for the data presented, and not as generalized statements on the effect of hot working. The properties of a piece of rolled or forged carbon steel, after the steel has cooled to atmospheric temperature, depend upon a large number of variables, any of which may have a marked effect. In none of the few published data have all of these variables been controlled or even considered by the investigator. The only fact, established without possibility of contradiction, is that, in steel containing the usual amount of other elements, hot worked at the usual temperatures, an increase in carbon content is accompanied by an increase in tensile strength, yield strength, and hardness and by a decrease in elongation, reduction of area, and impact resistance. Except for this one fact, nothing is known with exactness; the determination of the effect of all the variables in the manufacture and hot working of carbon steels is a wide-open field for research.

CHAPTER V

MECHANICAL PROPERTIES OF COLD-WORKED CARBON STEELS

Effect of Cold Working on Mechanical Properties—Properties of Cold-worked Steels of Varying Carbon Content—Effect of Prior Hot Working and Heat Treatment—Effect of Aging and Subsequent Heat Treatment—Author's Summary

Iron and steel, like other malleable metals and alloys whose recrystallization temperature occurs considerably above room temperature, can be hardened by cold work. Commercial iron-carbon alloys, ranging from the very low carbon grades to those containing as much as 1.5 per cent carbon, are cold rolled or cold drawn into sheet, strip, tubes, and wire of a variety of shapes and sizes. Cold working has two outstanding advantages: first and most important, certain sizes and shapes can be produced more readily by this means than by any other method; and second, properties can be attained by cold working which are difficult or even impossible to attain by any other commercial process.

For many years the processes of cold working have been the most empiric of all steel-making operations. Many of them have been plant secrets which were handed down from father to son for generations. The result has been that until recently there was little published information on cold-working processes and the effect of these processes on the properties. Within the past few years, however, the literature has increased at a remarkable rate so that the discussion in this chapter is not a correlation of all of the data; rather, it is necessarily a selection of values which show typical properties or which indicate the effect of the process on the properties.

A. EFFECT OF COLD WORKING ON MECHANICAL PROPERTIES

As is well known there are three major classes of cold-worked

the effect of cold working on the properties, the first is the most important. Most of the rolling or drawing operations in the commercial manufacture of sheet, strip, and tubes are carried on at elevated temperatures; ordinarily these products receive only a slight amount of cold reduction as a final operation, and frequently (usually, for sheet and strip) the effects of this cold work are destroyed by a subsequent annealing.

In the manufacture of wire, on the contrary, practically every mechanical working operation is performed cold; reductions in cross-section by drawing may be as great as 98 per cent. It was to be expected, therefore, that more data would be available which show the effect of cold working on the properties of wire than data which show the same effect on sheet, strip, and tubes. Although most of the published information on the latter products is discussed in this chapter, most of the work which is correlated below was done on wire.

50. The Operation of Cold Working.—Practically all of the commercial iron-carbon alloys which are cold rolled or drawn into industrial products are first hot worked into some form suitable for cold working. Ingots may be rolled into billets and billets to rods which are then cold drawn to wire. Ingots may also be rolled to slabs which are hot rolled further and then cold rolled into thin sheet and strip. Between the hot-working and the cold-working operations some form of heat treatment is occasionally used. Thus, for some grades of wire, the rod is "patented" before cold drawing by heating well above the critical temperature and cooling in air or in a lead bath maintained at a temperature of 500 to 600°C. (930 to 1110°F.). The operation of patenting and the advantages of the resulting fine pearlitic structure in wire drawing have been discussed in Volume I of this monograph (page 332).

The most pronounced effect of cold working is an increase in strength and hardness and a decrease in ductility as represented by elongation and reduction of area. The elongation may be reduced practically to zero—in other words, the elongation before rupture is almost entirely elastic; the reduction of area is usually decreased, but to a less marked degree. There is a limit to the amount of cold working which iron and steel will withstand without tearing apart internally. Before this limit is reached, the material must be heated to 400 to 650°C.

(750 to 1200°F.) to relieve internal strains, induce recrystallization, or both. This low-temperature or "process" annealing has been described in Volume I, page 283, of this monograph.

Summarizing the process of cold working briefly: commercial iron-carbon alloys may be processed into finished form by cold rolling or cold drawing a hot-worked section either directly; or after a structure-controlling heat treatment of the hot-rolled section known as patenting; or by cold working, process annealing, and further cold working; or by a combination of the above. In addition to these heat treatments, which are applied before the material is cold worked into final form, the finished wire or sheet may also be heat treated—annealed, or quenched and tempered—to produce any desired combination of properties within the limits possible for steel of that particular composition.

51. Mechanical Tests of Cold-worked Steel.—Before discussing the effect of cold working on the mechanical properties of steel a word in explanation of the terms used is desirable. Properties ordinarily determined on cold-rolled products are tensile strength and percentage elongation. The gage length for wire is usually 10 in., for sheet 2 in. Yield strength (frequently reported as yield point) and proportional limit are occasionally determined. Most cold-worked products are so thin or of such small diameter that the reduced area at the fracture can rarely be measured accurately. Hence, a value for the reduction of area of test pieces as a measure of the ductility of the steel seldom appears in the literature, and has not been so used in this chapter except in Tables 38 and 39 (pages 154 and 159). However, the term "reduction of area" in the usual wire-mill sense, signifying the amount of reduction by cold drawing, has been used frequently, and should not be confused with the same term used later in this chapter and in other parts of this monograph to denote one of the familiar mechanical properties.

Wire-mill laboratories frequently determine a property called "twist." This is obtained by clamping a section 6, 8, or 10 in. long in the jaws of a hand-operated machine; one jaw is fixed, the other revolves. The wire is twisted until it fails by shear; the number of complete revolutions of the movable head is the number of twists. Some authorities, for example Bonzel,[792] consider that this torsion test agrees closely with the reduction of

area as determined by the tensile test; others are not sure, and some go so far as to say that the only advantage of this simple test is that it may detect local defects in that section of the wire under test. Legge,* in supplying data quoted in the next section, stated: "We have not prepared curves showing the effect

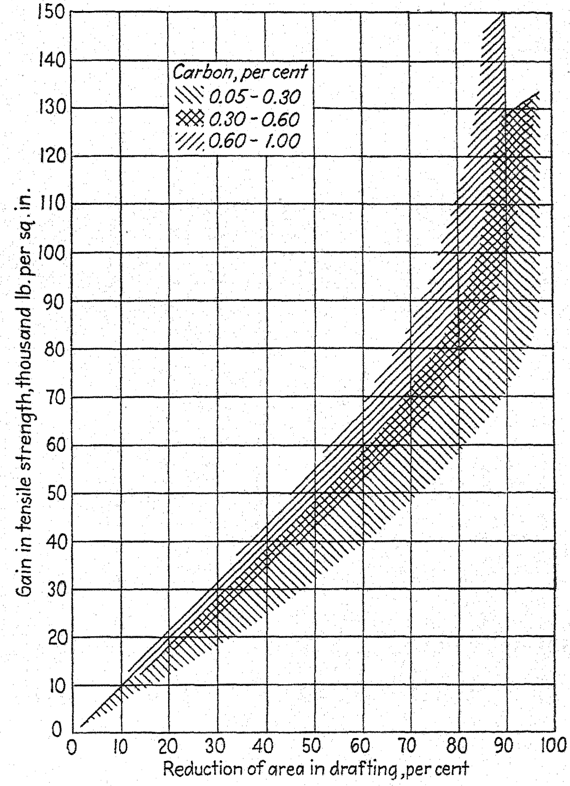

FIG. 21.—Effect of cold drawing on the tensile strength of wire containing 0.05 to 1.00 per cent carbon. (*Legge.*)

of drafting on twists because the results obtained have been very irregular . . . "

Some form of bend test is used almost universally in wire and strip mills, but often the test is made crudely and, although of great value in the hands of the experienced wire drawer, seldom gives results of more than qualitative interest.†

* Private communication, Mar. 13, 1935.

† An ingenious reverse bending machine for sheet and streamline wire,

52. General Effect of Drawing on Mechanical Properties.— Some interesting results on the effect of cold drawing on tensile strength and elongation in 10 in., of steels containing 0.05 to 1.00

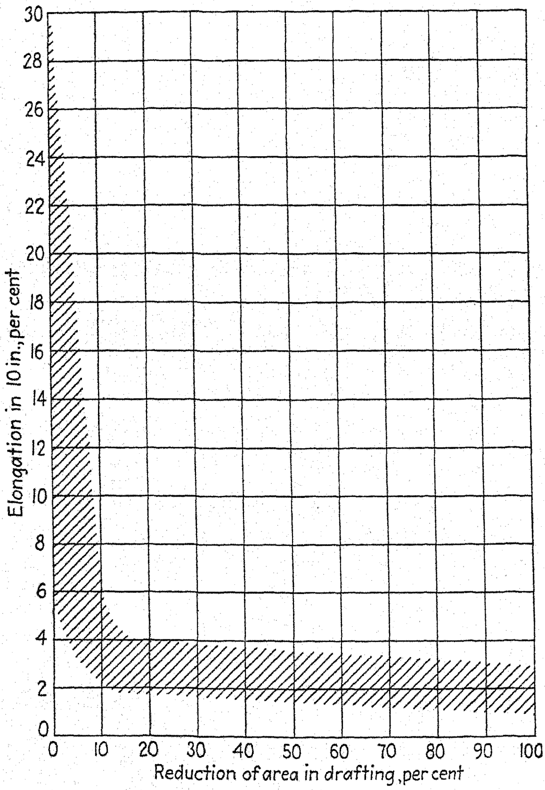

FIG. 22.—Effect of cold drawing on the elongation of wire containing 0.05 to 1.00 per cent carbon. (*Legge.*)

per cent carbon, are shown in Figs. 21 and 22. In the former, increase in tensile strength and, in the latter, percentage elongation are plotted against reduction of area in drawing. In supplying these charts, Legge* wrote:

In analyzing and plotting the data obtained for steels whose carbon content varied from 0.05 to 1.00 per cent, and whose original condition

which gives quantitative results, has been recently designed by D. M. Warner, testing engineer, U. S. Army Air Corps, Wright Field, Dayton, Ohio. See Warner, *Trans. Am. Soc. Mech. Eng.*, v. 54, 1932, p. AER 54-17-141.

* Private communication, Mar. 13, 1935.

varied from the green [untreated] rod to the double-lead-patented, it was found that in all the steels in the various conditions tensile strength increased in about the same manner. For this reason tensile gain, rather than ultimate tensile curves for all the steels in all the conditions, was plotted on the same chart.

On plotting the curves, it was observed that they grouped themselves according to three carbon ranges. Therefore, we have shown areas giving the most probable tensile strength developed for steels of these three carbon ranges. The exact curves were not given because it was found that the variation between two heats of the same type of steel was as great as that between two steels of considerable difference in carbon content, or in original condition. In general, it may be said that the higher the carbon, or the harder the original structure, the nearer the top of the range will be the tensile developed.

On the chart giving the change of elongation with drafting, we again plotted an area of the most probable elongation for all the steels in all the conditions, because it was found that irrespective of the original elongation, the elongation always dropped to about the same value after a 10 per cent reduction, and that further reduction lowered the elongation but slightly.

B. PROPERTIES OF COLD-WORKED STEELS OF VARYING CARBON CONTENT

A large proportion of the tonnage of wire made by the mills of the United States is the so-called "common grade." This is drawn from a hot-rolled, green (untreated) rod of low-carbon Bessemer or basic open-hearth steel; part of this production is used in the hard-drawn condition; a considerable proportion, however, is subjected to a final annealing, and/or is given a final corrosion-resisting coating of zinc or tin. Typical properties of common low-carbon wire after various amounts of cold working, together with some of the available data showing properties of cold-worked wire, sheet, strip, and tubes of varying carbon content, are discussed on the following pages.

53. Properties of Commercial Low-carbon Wire.—Mechanical properties typical of commercial low-carbon wire after varying amounts of cold work are given in Tables 30 to 32.* The analyses of the steels used for this grade of wire were as follows:

* Sisco, unpublished data.

Element	Composition, per cent	
	Bessemer steel	Basic open-hearth steel
Carbon	0.05 to 0.10	0.05 to 0.10
Manganese	0.35 to 0.45	0.35 to 0.45
Sulphur	0.12 or less	0.05 or less
Phosphorus	0.10 or less	0.04 or less

TABLE 30.—PROPERTIES OF HARD-DRAWN COMMERCIAL BESSEMER WIRE*

Rods		Drafts		Wire			
Gage†	Nominal diameter, in.	Number	Reduction in area, per cent	Gage†	Actual diameter, in.	Tensile strength, lb. per sq. in.	Elongation in 10 in., per cent
			Average (as hot rolled)			64,500	20.3
5/16	0.3125	1	17.8	1	0.2834	86,500	2.41
5/16	0.3125	1	29.3	2	0.2628	100,500	2.16
9/32	0.2813	1	24.8	3	0.2440	94,700	2.59
1/4	0.250	1	19.4	4	0.2244	90,950	2.76
1/4	0.250	1	31.8	5	0.2065	108,650	1.66
4	0.2253	1	26.8	6	0.1928	101,400	2.08
4	0.2253	1	37.3	7	0.1784	109,800	1.51
5	0.2070	1	39.7	8	0.1607	108,850	1.32
5	0.2070	1	44.3	8.5	0.1546	109,300	1.36
5	0.2070	2	48.9	9	0.1480	112,250	1.11
5	0.2070	2	56.4	10	0.1351	120,750	1.13
5	0.2070	2	59.7	10.5	0.1286	116,250	1.15
5	0.2070	2	65.0	11	0.1210	117,600	0.94
5	0.2070	2	73.8	12	0.1044	124,100	0.80
5	0.2070	3	78.3	12.5	0.0976	130,700	0.44
5	0.2070	3	80.1	13	0.0909	141,600	0.47
5	0.2070	4	84.8	14	0.0800	146,250	0.65
5	0.2070	4	87.2	15	0.0720	155,500	0.35
5	0.2070	5	91.2	16	0.0607	158,600	0.35

* Sisco.

† American Steel and Wire Company gage (old Washburn and Moen).

Table 30 shows the effect of one to five drafts (20 to 90 per cent reduction in area) on the tensile strength and elongation. The values for the hot-rolled rod are averages of nine specimens; the values for the wire are actual determinations and are typical of wire of this composition, drawn the amount shown from rod of five sizes. As the diameters given in Table 30 are nominal for the rod and actual for the wire, the calculated reduction is only accurate to about 0.2 per cent.

TABLE 31.—PROPERTIES OF COMMERCIAL BESSEMER WIRE ANNEALED AND DRAWN ONE DRAFT AFTER ANNEALING*

Gage†	Actual diameter, in.	Annealed		One draft after annealing	
		Tensile strength, lb. per sq. in.	Elongation in 10 in., per cent	Tensile strength, lb. per sq. in.	Elongation in 10 in., per cent
1	0.2824	58,900	17.6		
2	0.2620	61,300	23.0		
3	0.2442	64,100	22.5		
4	0.2283	61,600	16.9	81,300	2.11
5	0.2062	66,600	18.5	79,400	2.08
6	0.1939	61,800	20.4	79,500	1.65
7	0.1758	57,500	20.0	87,350	1.45
8	0.1615	59,100	20.1	93,500	1.23
9	0.1436	69,450	20.5	94,700	1.36
10	0.1340	58,800	21.8	91,900	1.65
11	0.1187	61,600	19.2		
12	0.1037	66,900	22.5		
13	0.0916	70,900	22.8		
14	0.0788	69,750	23.7		
15	0.0712	69,600	18.8		
16	0.0625	72,500	19.5		

* Sisco.
† American Steel and Wire Company gage (old Washburn and Moen).

Table 31 gives the properties of Bessemer wire of various gages as annealed and as drawn one draft after annealing. It will be noted from both of these tables that one draft is enough to reduce the elongation to about 2 per cent or even less.

The properties of three grades of basic wire are given in Table 32. This wire was drawn from a green rod which had a tensile strength (average of 17 specimens) of 54,500 lb. per sq. in. and

TABLE 32.—DRAFTING PRACTICE AND PROPERTIES OF THREE GRADES OF BASIC WIRE*

Final size of wire		Basic			Soft basic				Annealed basic	
Gage†	Nominal diameter, in.	Number of drafts	Tensile strength, lb. per sq. in.	Elongation in 10 in., per cent	Number of drafts: Before annealing	Number of drafts: After annealing	Tensile strength, lb. per sq. in.	Elongation in 10 in., per cent	Tensile strength, lb. per sq. in.	Elongation in 10 in., per cent
2	0.2625	1	82,800	2.05	..	..	......	....	46,600	28.1
3	0.2437	1	88,450	2.35	..	..	......	....	47,600	27.8
4	0.2253	1	83,300	1.25	1	1	55,000	5.12	50,450	29.7
5	0.2070	1	80,100	1.25	1	1	60,500	4.75	48,300	23.9
6	0.1920	1	89,600	1.75	1	1	71,100	5.70	50,600	24.1
7	0.1770	1	88,600	1.50	2	1	68,600	4.50	47,600	24.7
8	0.1620	1	88,150	1.50	2	1	61,500	3.55	43,600	27.6
9	0.1483	2	89,200	1.50	2	1	64,800	2.75	52,000	22.3
10	0.1350	2	93,950	1.26	2	1	65,900	3.00	51,800	28.1
11	0.1205	2	98,800	1.00	2	1	66,100	2.75	47,900	20.4
12	0.1055	3	100,800	1.50	3	1	68,500	1.25	44,900	25.0
13	0.0915	3	111,550	1.25	3	1	78,500	1.25	44,800	20.5
14	0.0800	4	102,400	1.25	4	1‡	84,800	1.00	54,300	20.0
15	0.0720	4	110,500	1.00	4	1‡	85,300	1.25	50,500	23.0
16	0.0625	5	122,100	1.00	5	1‡	77,100	1.50	54,650	18.0
17	0.0540	5	128,200	1.50	3	2‡	78,300	1.00	54,100	21.0
18	0.0475	6	125,700	1.50	3	3‡	79,300	1.25	57,700	20.0

* Sisco.
† American Steel and Wire Company gage (old Washburn and Moen).
‡ On sizes 14 to 18 the wire was reannealed after the drafting shown in this column and was drawn one draft after the second annealing.

an elongation in 10 in. of 24.9 per cent. The rods were drawn the number of drafts shown to wire of the gage given at the left of the table. The "basic" wire was drawn one to six drafts without annealing and had a tensile strength in the finer sizes up to 125,000 lb. per sq. in. The "soft basic" wire was drawn one to five drafts, annealed, and drawn one draft after annealing. For

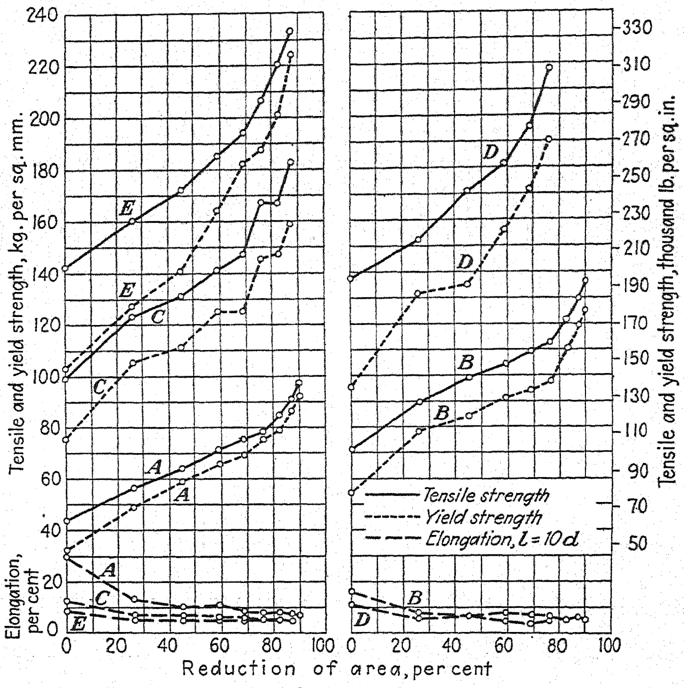

FIG. 23.—Effect of the amount of drafting on the tensile properties of wire containing 0.10 to 1.01 per cent carbon. (*Eicken and Heidenhain.*[116])

the finer sizes (14 to 18 gage) the material was drawn, annealed, drawn again, and reannealed. In every case, however, the wire received only one draft after the final annealing, which resulted in 65,000 to 85,000 lb. per sq. in. tensile strength for all but the larger sizes. The elongation of the soft basic wire is but slightly higher in the larger sizes and no higher in the finer sizes than that of the hard-drawn basic wire.

The properties of the "annealed basic" show that a final annealing with its accompanying recrystallization destroys most

or in some cases all of the effects of cold work, and restores most or all of the original ductility; in fact, the elongation of some of the specimens of the annealed wire is somewhat higher and the tensile strength somewhat lower than the corresponding properties of the rod from which it was drawn.

As will be noted from these tables, the tensile strength of unannealed wire depends not only upon the number of drafts but also upon the amount of reduction per draft. Thus, the first specimen of wire in Table 30, drawn one draft (17.8 per cent reduction) from a 5⁄16-in. rod, had a tensile strength of 86,500 lb. per sq. in., while the second, drawn one draft (29.3 per cent reduction) from a rod of the same size, had a strength of 100,500 lb. per sq. in.

54. Properties of Cold-drawn Wire and Rod of Varying Carbon Content.—Adam(54) and Bonzel(792) have given a few data on the mechanical properties of wire of varying carbon content. Although these are of interest, the variation in carbon of the specimens tested is not wide enough to warrant reproduction of the data. In general, the change in tensile strength with drafting corresponds to that shown in Legge's curves given in Fig. 20.

In work reported by Eicken and Heidenhain(116) on cold-drawn wire, steels of the following composition were used:

Steel	Composition, per cent				
	C	Mn	Si	S	P
A	0.10	0.54	0.06	0.023	0.012
B	0.26	0.76	0.20	0.033	0.031
C	0.54	0.56	0.25	0.024	0.022
D	0.83	0.61	0.32	0.035	0.024
E	1.01	0.72	0.18	0.027	0.025

The rods were 5 mm. (0.20 in.) in diameter and were patented by quenching in a lead bath. They were then cold drawn with different reductions per pass and with varying speeds. Results of drawing with an average reduction of 25 per cent per pass are shown in Fig. 23 where the properties are plotted against total reduction in cross-sectional area, and where the plotting points

represent a constant reduction per pass of 25 per cent.* The elongation was measured on a gage length equal to 10 times the diameter of the wire. The curves representing tensile and yield strengths (Fig. 23) are approximately parallel; the strength for any amount of reduction increased as the carbon increased. Experiments with different drawing speeds and different reductions per pass indicated that the drawing speed had no influence on the properties of the wire but that the reduction per pass did have a measurable effect. This is discussed in detail in section 57.

In studying the influence of cold work on the properties of carbon and alloy steels, Greulich[373] used five carbon steels of the following analyses:

Number	Composition, per cent				
	C	Mn	Si	S	P
1	0.12	0.40	0.00	0.024	0.017
2	0.29	0.40	0.06	0.013	0.020
3	0.34	0.57	0.18	0.017	0.024
4	0.46	0.75	0.30	0.028	0.021
5	0.61	0.83	0.34	0.016	0.022

Square bars were annealed for 3 hr. at 650°C. (1200°F.) and cold rolled to square sections with reductions in area up to 50 or 55 per cent. The tensile properties and hardness numbers for steels 1, 2, 4, and 5 versus reduction in cross-sectional area are plotted in Fig. 24. The curves for the medium-carbon steel, No. 3, which are not reproduced, were almost identical with the curves for steel 2, differing only to the degree expected for the slight increase in carbon content.

As will be noted from Fig. 24, the curves for the various steels are almost parallel, thus indicating that the increase in strength and hardness and the decrease in ductility with higher reductions in area are practically independent of the carbon content.

* Nead (private communication) commented that the curves in Fig. 23 show a tendency toward infinite values with increasing cold reduction, and that this is contrary to his experience and to the data of other investigators quoted later in this chapter.

Greulich arrived at a similar conclusion by replotting the curves shown in Fig. 24 on a logarithmic scale and deriving from the resulting curves an equation in which the constant b represents the strengthening by cold work. Taking the yield strength as the most accurate measure of this strengthening he found that b is the same for all cold-worked carbon steels, thus indicating

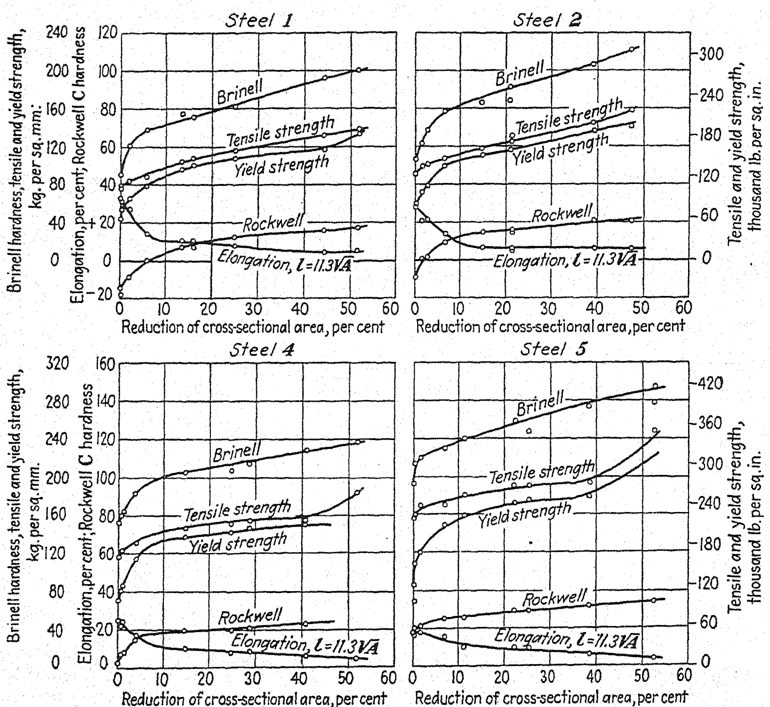

Fig. 24.—Effect of the amount of reduction in cold rolling on the tensile properties of rod containing 0.12 to 0.61 per cent carbon. (*Greulich.*[373])

that the increase in strength by cold work is independent of the carbon content.

Properties of a series of steels tested by Altpeter as quoted by Bonzel[792] are shown in Fig. 25 in which tensile strength is plotted against carbon content. Here again—if allowance is made for differences in the kind of steel, minor variations in composition, and in wire-drawing practice and possible slight variations in the air patenting (different cooling rates, changes in critical temperature with carbon, etc.)—drafting is shown to

increase the tensile strength by an amount which is apparently independent of the carbon content.

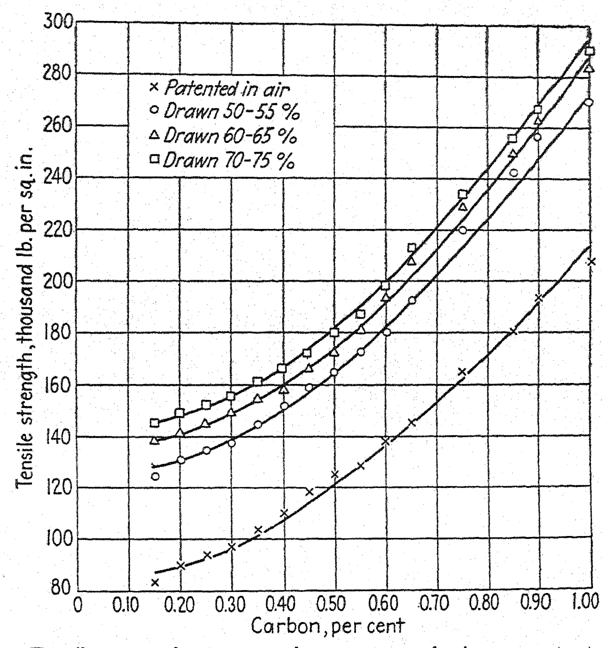

FIG. 25.—Tensile strength versus carbon content of wire, as patented and as cold drawn. (*Bonzel*[792] *after Altpeter.*)

55. Properties of Cold-rolled Strip of Varying Carbon Content. Swinden and Bolsover[239] determined the effect of cold rolling on the properties of hot-rolled strip steel of the following analyses:

Steel	Composition, per cent				
	C	Mn	Si	S	P
A	0.10	0.41	Trace	0.036	0.017
B	0.34	0.72	0.095	0.037	0.030
C	0.54	0.78	0.105	0.032	0.036
D	0.70	0.68	0.101	0.030	0.028

According to the investigators, the results are representative of those obtained in actual practice. In addition to tensile strength and percentage elongation, the proof stress (the elastic

extension plus 0.1 per cent of the gage length), Brinell hardness, and shear strength were also determined.

The effect of cold working, with reductions in thickness up to 67 per cent, on tensile strength and elongation is shown by the values given in Table 33. The effect of the various reductions and of the carbon content is clearly indicated. Proof-stress values were rapidly increased by cold rolling with a 7 per cent reduction in thickness and then, for all higher reductions, followed

TABLE 33.—EFFECT OF COLD ROLLING ON TENSILE STRENGTH AND ELONGATION OF STRIP*

Nominal size, in.	Reduction of thickness, per cent	Steel *A* (0.10 % carbon)		Steel *B* (0.34 % carbon)		Steel *C* (0.54 % carbon)		Steel *D* (0.70 % carbon)	
		Tensile strength, lb. per sq. in.	Elongation in $4\sqrt{A}$, per cent	Tensile strength, lb. per sq. in.	Elongation in $4\sqrt{A}$, per cent	Tensile strength, lb. per sq. in.	Elongation in $4\sqrt{A}$, per cent	Tensile strength, lb. per sq. in.	Elongation in $4\sqrt{A}$, per cent
0.300	..	61,600	31.1	93,000	27.8	111,800	23.4	128,900	22.1
0.280	7	72,000	22.2	108,800	16.0	120,700	17.0	138,400	16.7
0.260	13	75,000	19.4	117,600	13.5	128,100	12.5	150,100	12.0
0.240	20	80,800	17.6	117,400	11.0	135,600	9.1	158,000	12.4
0.220	26	84,000	16.3	124,300	10.8	142,200	9.4	160,300	10.7
0.200	33	83,800	15.0	126,300	10.0	145,800	8.4	163,200	11.4
0.180	40	89,400	13.0	132,900	8.2	154,800	7.1	173,000	8.6
0.160	47	90,900	11.3	138,100	9.6	159,500	8.9	176,400	11.1
0.140	53	96,700	9.2	143,400	8.5	164,000	7.7	186,800	10.0
0.120	60	101,900	8.7	146,500	7.9	173,600	6.7	195,500	8.2
0.100	67	103,000	8.7	154,100	8.2	182,000	6.4	198,300	8.2

* Swinden and Bolsover.(239)

a curve approximately parallel to the tensile strength. Brinell hardness was increased by cold work, but the values were variable and did not bear such a definite relation to tensile strength as do the Brinell values of material which has not been cold worked. Shear stress was increased uniformly but slightly by cold working. For example, by reductions in thickness of 67 per cent the shear stress of the 0.54 per cent carbon material increased from 78,600 to 97,700 lb. per sq. in. (24 per cent), while the tensile strength increased from 111,800 to 182,000 lb. per sq. in. (62 per cent).

Swinden and Bolsover also determined the effect of cold rolling on the properties of sheet which had been hot rolled and normalized or annealed. This is discussed briefly in a later section (section 60).

The effect of cold rolling on the properties of strip of different carbon contents was also investigated by Pomp and Poellein.[331] The steels used were of the following analyses:

Number	Composition, per cent				
	C	Mn	Si	S	P
1	0.09	0.36	0.03	0.052	0.021
3	0.29	0.28	0.05	0.005	0.019
6	0.65	0.28	0.04	0.032	0.019
9	0.93	0.26	0.04	0.023	0.021
12	1.14	0.26	0.03	0.026	0.020
14	1.44	0.25	0.04	0.019	0.020

Sections 1.50 mm. (0.059 in.) thick were heat treated to produce lamellar pearlite, spheroidized cementite, or a sorbitic (fine pearlitic) structure and then cold rolled with different reductions. The tensile strength and the elongation for a 50-mm. (1.97-in.) gage length of the strips cold rolled from the spheroidized material are shown in Fig. 26 as a function of the thickness of the strip and the total reduction by cold rolling. With these specimens, as with the wire and rod discussed in the previous section, the increase in tensile strength and the decrease in elongation were almost wholly dependent upon the amount of cold working and apparently independent of the carbon percentage. Pomp and Poellein also determined the Herbert "time hardness" of the six steels. The hardness value increased with the amount of reduction in cold working along curves very similar in slope to the curves for tensile strength given in Fig. 26.

56. Properties of Cold-drawn Seamless Tubes of Varying Carbon Content.—The effect of cold working on the properties of seamless tubing was determined by Pomp and Albert[229] using basic open-hearth steels of the composition shown in the table on page 127.

The tubes were hot rolled and then cold drawn in two drafts, with intermediate annealing, to 70 mm. (2.756 in.) outside diam-

eter and 8 mm. (0.315 in.) wall thickness. Three series of specimens were subjected to the following treatments:

A. Annealed 30 min. slightly above A_3 and cooled in air.

B. Heated 30 min. slightly above A_3 and quenched: steels 1, 2, 3, and 4 in water; steel 5 in oil; all were then tempered 30 min. at 650°C. (1200°F.).

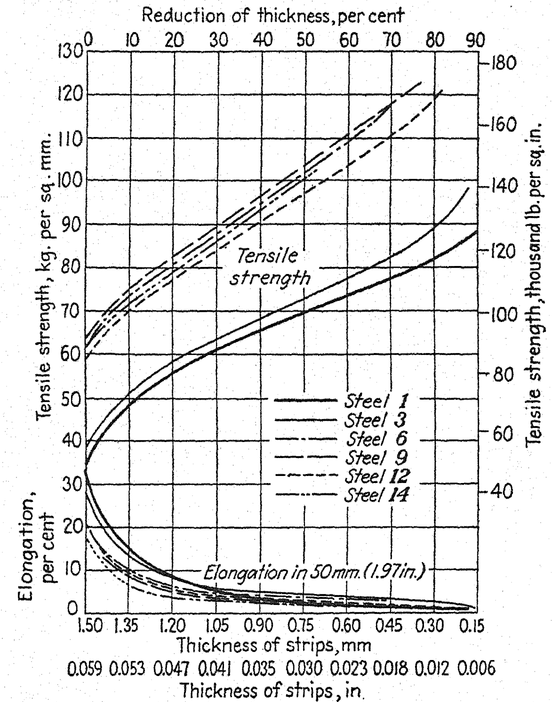

FIG. 26.—Effect of the amount of reduction in cold rolling on the tensile properties of strip containing 0.09 to 1.44 per cent carbon. The composition of the steels used is given on page 125. (*Pomp and Poellein.*[331])

C. Patented by heating 30 min. slightly above A_3 and quenched in molten lead at 500 to 520°C. (930 to 970°F.).

Specimens representing each of the above treatments were cold worked with from 8 to 50 or 60 per cent reduction by three methods; by the first, called "pressure drawing," the specimens were drawn without a plug; by the second, called "plug drawing," the specimens were drawn with a short internal plug; by the third, called "mandrel drawing," the specimens were drawn over a mandrel. In the first, pressure drawing, the cross-sectional

reduction results from decreasing the tube diameter while maintaining the same wall thickness. In plug drawing the reduction results from a decrease in both diameter and wall thickness. This is also true of mandrel drawing.

Impact tests were made on a Charpy 45-deg. V-notch specimen, 120 mm. (4.724 in.) long, and 12 mm. (0.472 in.) thick. It was tested on a 90-mm. (3.543-in.) span. The width was variable, being the wall thickness of the tube. The specimen was cut longitudinally and the curvature of the tube was not machined off; it was therefore of unusual shape as well as being variable in width. Impact values are given in m-kg. per sq. cm. and were calculated to the area back of the notch for each size of specimen. The Brinell-hardness determinations were made with a 5-mm. ball and a 750-kg. load.

The effect of cold working by pressure drawing on specimens subjected to treatments *A*, *B*, and *C* is shown in Fig. 27; for cold working by plug drawing in Fig. 28, and for mandrel drawing in Fig. 29. For each steel, each heat treatment, and each method of cold working, tensile strength increased with the amount of reduction and, with a few exceptions, was almost independent of the carbon content. As is also the case in wire drawing elongation dropped sharply with the first application of cold work, and then decreased but slightly with increased reduction. In general, Brinell-hardness values increased along curves similar to the tensile strength.

Because of the unusual specimens used the only interpretation of the impact results is that the resistance of all of the steels,

Number	Composition, per cent				
	C	Si	Mn	S	P
1	0.10	0.14	0.54	0.028	0.022
2	0.28	0.29	0.59	0.045	0.024
3	0.41	0.32	0.54	0.025	0.028
4	0.57	0.09	0.45	0.031	0.017
5	0.84	0.14	0.32	0.017	0.014

except the 0.10 per cent carbon material, was low for reductions of more than 10 per cent.

57. Effect of Variations in Cold-working Practice on Properties.—As is evident from the data given in the preceding sections,

the principal factor affecting the mechanical properties of cold-worked steel is the amount of cold working, in other words the amount of reduction in cross-sectional area by cold drawing or rolling, either directly from the hot-rolled section or after a preliminary or an intermediate heat treatment. As a general rule, it may also be said that the increase in tensile strength caused by cold work has apparently no close relation to the carbon content. However, there are other factors, in addition to the

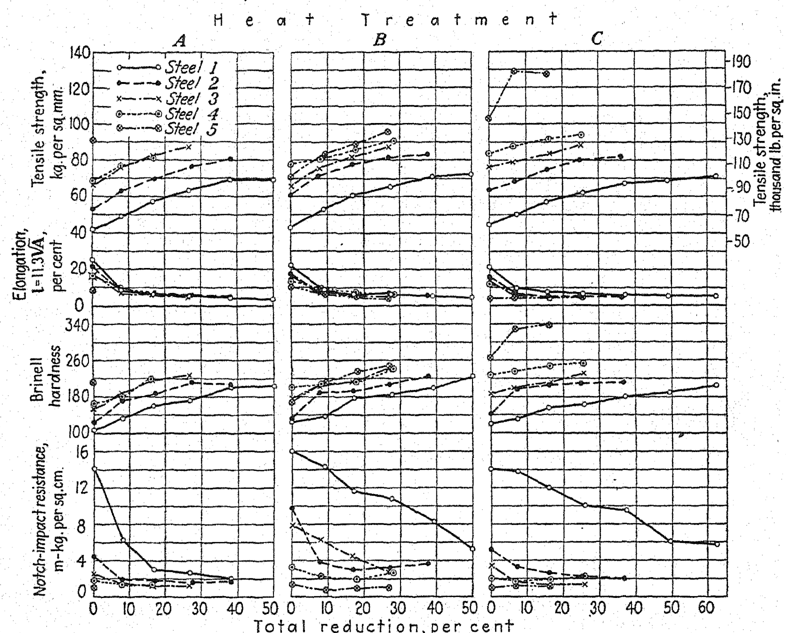

FIG. 27.—Effect of the amount of reduction in pressure drawing on the properties of tubes containing 0.10 to 0.84 per cent carbon. (*Pomp and Albert.*[229])

amount of reduction in cross-sectional area, which may affect the properties. One of these, as stated on page 120, is the amount of reduction per draft or pass.

Eicken and Heidenhain,[116] in addition to investigating the properties of cold-drawn wire of various carbon percentages (see Fig. 23), also determined the effect of the amount of reduction per draft. Three of their steels, *A*, *C*, and *E*, whose composition is given on page 120, were drawn four and six drafts from 5 mm. (0.197 in.) to 2.1 mm. (0.083 in.). The total reduction in area was the same, but the amount of reduction per draft was, of

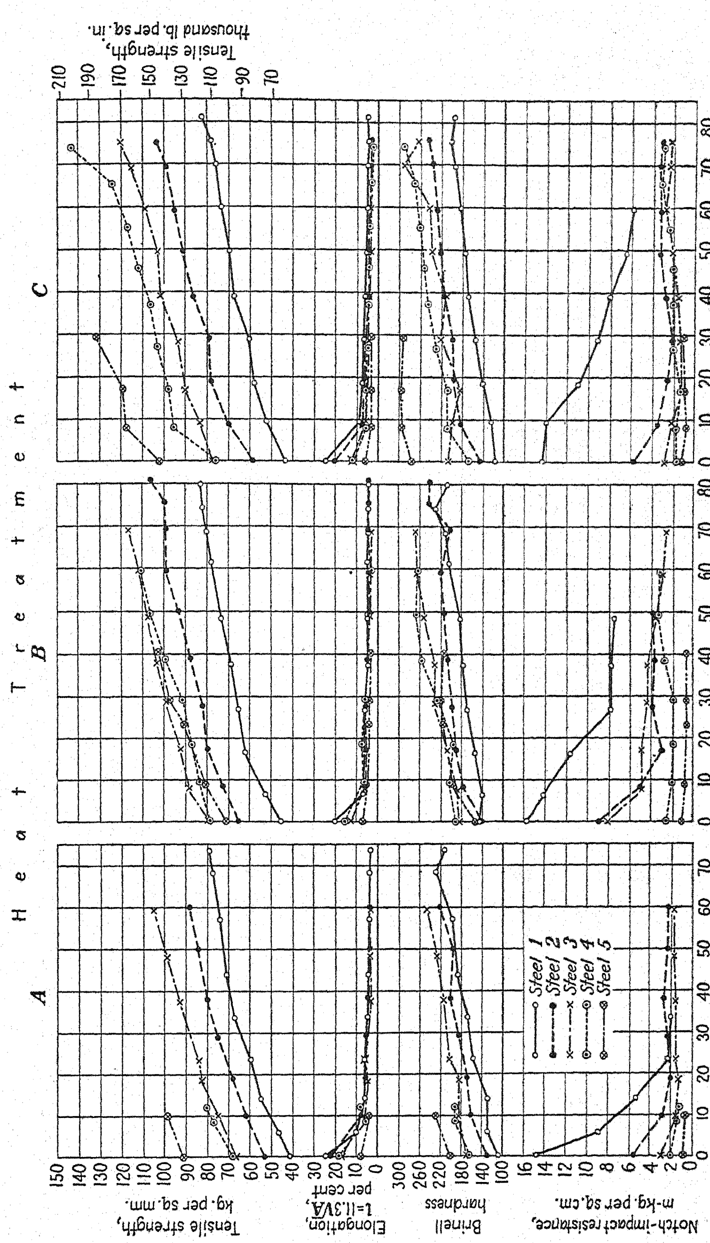

FIG. 28.—Effect of the amount of reduction in plug drawing on the properties of tubes containing 0.10 to 0.84 per cent carbon. (*Pomp and Albert.*[229])

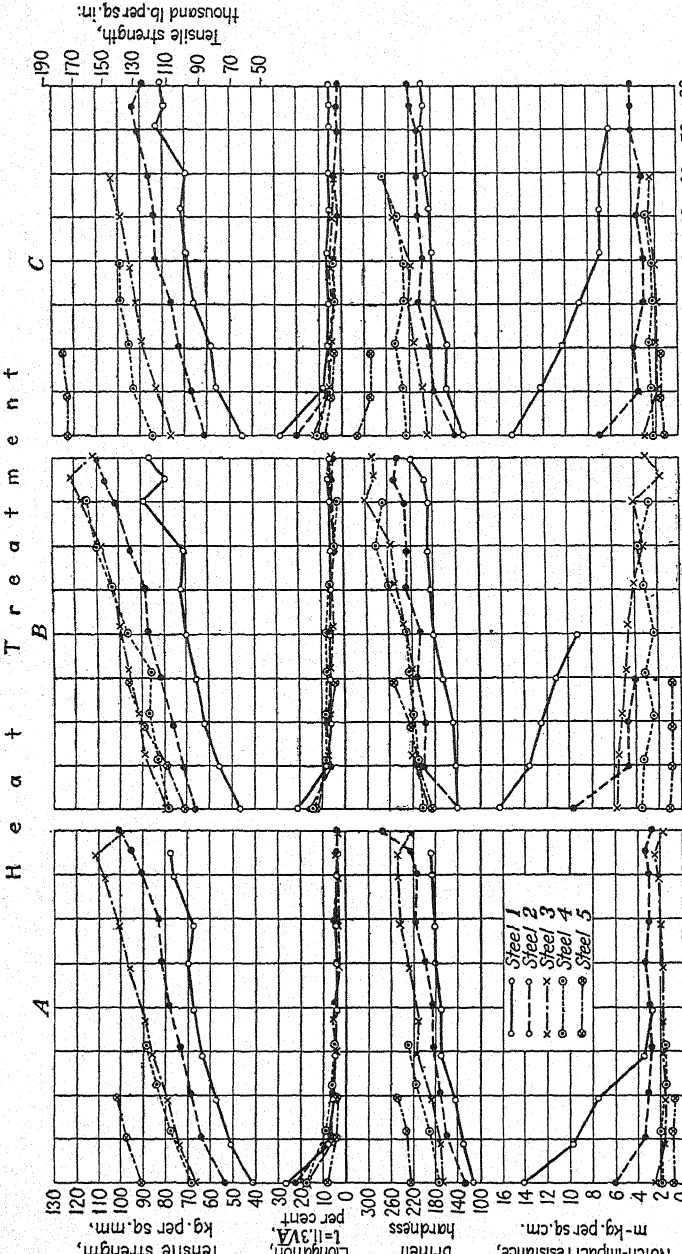

FIG. 29.—Effect of the amount of reduction in mandrel drawing on the properties of tubes containing 0.10 to 0.84 per cent carbon. (*Pomp and Albert*.[229])

course, heavier for the wires drawn four drafts than for those drawn six drafts. Tensile and yield strengths are given in Table 34.

TABLE 34.—EFFECT OF REDUCTION PER DRAFT ON TENSILE PROPERTIES*

Steel	Carbon, per cent	Number of drafts	Tensile strength, lb. per sq. in.	Yield strength, lb. per sq. in.
A	0.10	4	123,700	112,400
A_1	0.10	6	120,900	109,500
C	0.54	4	256,000	220,500
C_1	0.54	6	230,400	206,200
E	1.01	4	351,300	341,400
E_1	1.01	6	314,300	285,900

* Eicken and Heidenhain.[116]

Bonzel[792] also gave some values for properties of steels drawn with heavy and light reductions per draft. The composition of the material used was not given; probably it was low-carbon steel. One lot of specimens was drawn from 0.200 to 0.140-in. wire in five and two drafts; another lot from 0.200 to 0.060-in. wire in sixteen and six drafts. The values obtained were as follows:

Material	Number of drafts	Tensile strength, lb. per sq. in.
0.140-in. wire..............	5	82,000
	2	87,000
0.060-in. wire..............	16	115,000
	6	128,000

These results confirm the properties obtained by Eicken and Heidenhain (Table 34) and indicate that heavy drafts increase the tensile and yield strengths more than light drafts for the same total reduction.

Regarding the effect of some of the other variables in wire-drawing practice, Bonzel[792] stated: "We have never been able to

trace any effect of lubricant, die shape, or drawing speed on the properties of the wire drawn, and in this matter we are in accord with Giraud[254] who has studied the effect of these factors in the drawing of mild steel." Eicken and Heidenhain also found that (see page 121, section 54) drawing speed had no effect on the properties of steels containing 0.10 to 1.01 per cent carbon.

Ellinger* does not agree with the conclusions of Bonzel, Giraud, and Eicken and Heidenhain, that these variables have no effect on the properties: "This may be true in drawing mild steel, but each of these factors does have a definite effect on the properties in the case of high-quality wire."

Owing to the taper of the die walls, the flow of the metal in drawing a wire is not the same at the surface as at the center. To discover if changing the direction of drawing had any effect on the properties, Bonzel[792] divided a bundle of low-carbon steel rod, drew one half of the bundle several drafts in the same direction, and drew the other half by alternating the direction at each draft. All other variables were constant. Upon determining the properties after the drafts, no difference in tensile, bend, or torsion test could be detected.

To determine the effect of direction of working on properties of strip, Phillips and Dunkle[811] studied an open-hearth steel of 0.08 per cent carbon, 0.38 per cent manganese, 0.032 per cent sulphur, and 0.008 per cent phosphorus (silicon content not stated, the steel was evidently a rimmed steel free from silicon) after varying amounts of hot and cold rolling and after annealing at various temperatures. Tests were made promptly after rolling in order to avoid the complicating effect of age hardening. While these authors did not discuss the cleanness of the steel, they consider that their data indicate "crystallographic fibering," the effect of orientation of the crystals, rather than that due to elongated bands of inclusions and inhomogeneities.

The material was hot rolled to the four gages shown in Fig. 30(*a*), the finishing temperatures being 730, 775, 800, and 830°C. (1345, 1425, 1470, and 1525°F.), the temperature being the lower the lighter the gage. Samples were taken longitudinally and at an angle of 22.5, 45, 67.5, and 90 deg. from the direction of rolling on hot-rolled, cold-rolled, and annealed materials.

* G. A. Ellinger, Metallurgical Division, National Bureau of Standards, private communication.

These four gages of stock were then cold rolled to the same thickness, 0.032 in., which represents a reduction of 40, 50, 60, and 69 per cent. The tensile strength in the 90-deg. direction rises about 10 per cent above that in the longitudinal, and this holds irrespective of the percentage of reduction. The elongation in 2 in., not plotted in Fig. 30(*b*), varied from 2 to 3 per cent at different orientations for the material reduced 40 per cent, 1.5 to 2 per cent for that reduced 50 per cent, 1.75 to 2 per cent

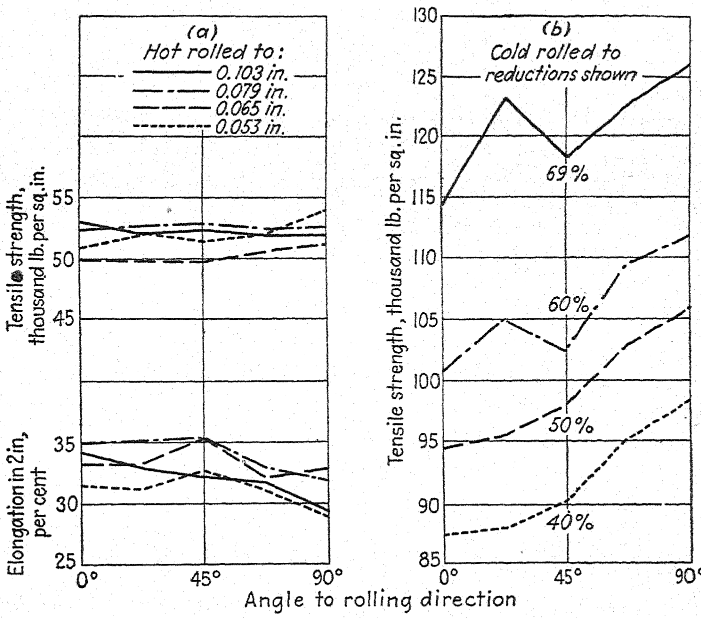

FIG. 30.—Effect of the direction of working on the tensile properties of hot- and cold-rolled strip. (*Phillips and Dunkle.*[811])

for that reduced 60 per cent, and 1.5 to 2 per cent for that reduced 69 per cent, with no indication that the orientation played a major part in these slight variations.

The four cold-rolled lots were then each annealed at 680, 705, 720, and 745°C. (1255, 1300, 1330, and 1375°F.), being held at temperature 2 hr., and furnace cooled. These 16 lots were each tested at the five orientations to the rolling direction. The results are shown in Fig. 31. Cupping tests showed the characteristic "ears" that result from grain orientation. On the material that had had 40 per cent reduction the ears were at

45 deg. to the rolling direction. On those with higher reduction, they were at 0 and 90 deg., the height of the ear increasing with the reduction.

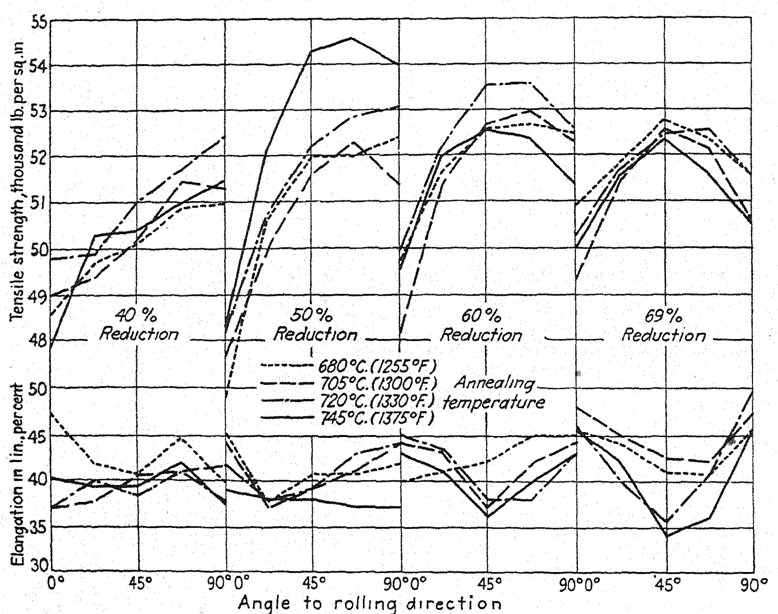

FIG. 31.—Effect of the direction of working and of annealing on the tensile properties of hot- and cold-rolled strip. (*Phillips and Dunkle.*[811])

C. EFFECT OF PRIOR HOT WORKING AND HEAT TREATMENT

The discussion of mechanical properties in the preceding pages was confined, so far as possible, to: (*a*) the effect of cold working on steels of varying carbon content including typical properties of wire, rod, tubes, and strip, (*b*) the effect of amount of reduction in cold working on properties, and (*c*) the effect of variations in cold-working practice. In this discussion it was impossible wholly to separate the properties of material which was drawn or cold rolled from hot-worked carbon steel from the properties of material which was cold worked after some prior or intermediate heat treatment. Thus, Tables 31 and 32 give properties of low-carbon Bessemer and basic wires which were annealed and drawn after annealing. Furthermore, in most of the other investigations quoted (see results of Eicken and Heidenhain,[116]

Greulich,[373] Pomp and Poellein,[331] and Pomp and Albert[229]) the material had been heat treated before cold working (for medium- und high-carbon steels this may be absolutely necessary so that the material can be cold worked to the required degree).

Low-carbon steels are almost never subjected to a preliminary heat treatment. They are, however, frequently annealed during cold working to restore the ductility so that they may be cold drawn or rolled to smaller sizes without failure, or so that specified properties may be attained. An illustration of the effect of this intermediate annealing on low-carbon wire is given in Tables 31 and 32. Medium- and high-carbon steels are frequently heat treated before cold working; in fact, patenting of some kind is usual for spring and other high-grade wires.

In the discussion which follows, an effort has been made to show the effect of variations in the prior hot-rolling practice, or the effect of prior or intermediate heat treatment, on the properties of the cold-worked material; attention has been paid chiefly to those investigations in which the prior mechanical or thermal treatment is the only or at least the most important variable.

58. Effect of Variations in Hot-rolling Practice on the Properties of Low-carbon Strip.—In their investigation of the effect of mill and annealing practices on the properties of sheet for deep drawing, Nead, Mahlie, and Dittrich* studied the properties of material after cold rolling with reductions of 2 to 65 per cent, and after subjecting the cold-rolled sheet to a variety of annealing and normalizing treatments. The object of the investigation was to work out the rolling and thermal treatments which would produce the most ductile sheet for deep drawing; hence, in their conclusions they considered not only the tensile properties but also the Rockwell *B* hardness and the results of the Olsen ductility tests. Results of a study of the microstructure also affected their final conclusions.

The steel used was a 0.05 per cent carbon, 0.37 per cent manganese basic open-hearth material, the complete analysis of which is given on page 102, where details of the finishing temperatures in rolling and the annealing practice are also given.

The effect of cold rolling on the tensile properties of strip, flat rolled with finishing temperatures of 730, 790, and 880°C. (1345,

* See footnote, p. 102.

1450, and 1615°F.), is shown in Fig. 32. It will be noted that the yield-strength curve joins the tensile-strength curve at reductions of 20 to 25 per cent. The elongation values drop to about 2 per cent with reductions of 35 per cent and remain practically constant as the reduction increases. A finishing temperature of 880°C. (1615°F.) results in higher tensile-strength values

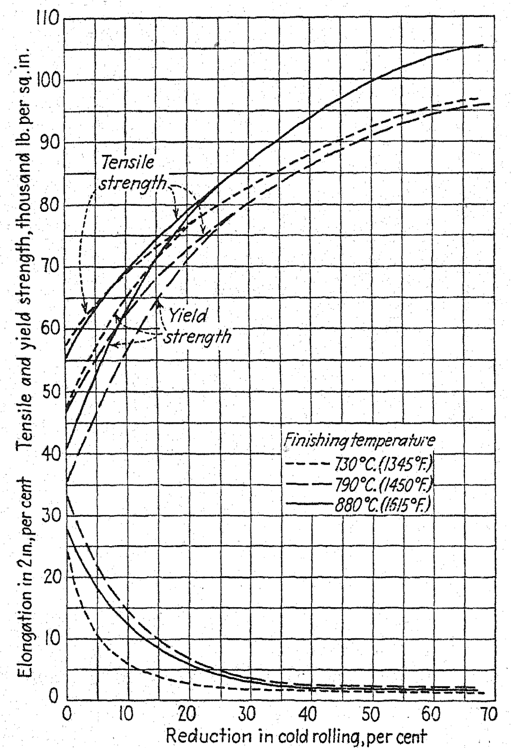

FIG. 32.—Effect of cold rolling on tensile and yield strengths of 0.08-in. strip rolled with finishing temperatures of 730, 790, and 880°C. (1345, 1450, and 1615°F.). (*Nead and associates.*)

for reductions of 25 per cent or more than the lower finishing temperatures.

The effect of finishing temperatures in hot rolling on the properties of sheets after a variety of annealing and normalizing treatments is shown in Table 35 for strip cold rolled with a reduction of 50 per cent. It may be concluded from these data that, in general, the best properties result from rolling flat with finishing temperatures of 790°C. (1450°F.) or above. Coiling the

TABLE 35.—EFFECT OF HOT-WORKING TEMPERATURES ON THE PROPERTIES OF LOW-CARBON STRIP 0.08 IN. THICK, AFTER COLD ROLLING WITH A REDUCTION OF 50 PER CENT, AND AFTER COLD ROLLING AND ANNEALING*

Annealing temperature		Finishing temperature											
		730°C. (1345°F.), flat			790°C. (1450°F.), flat			790°C. (1450°F.), coiled			880°C. (1615°F.), flat		
°C.	°F.	Tensile strength, lb. per sq. in.	Yield strength, lb. per sq. in.	Elongation in 2 in., per cent	Tensile strength, lb. per sq. in.	Yield strength, lb. per sq. in.	Elongation in 2 in., per cent	Tensile strength, lb. per sq. in.	Yield strength, lb. per sq. in.	Elongation in 2 in., per cent	Tensile strength, lb. per sq. in.	Yield strength, lb. per sq. in.	Elongation in 2 in., per cent
Cold rolled		94,100	94,100	1.5	91,400	91,400	2.0	87,400	87,400	2.5	102,900	102,900	2.5
595	1100	48,100	29,200	33.0	44,400	29,100	35.0	45,100	29,700	30.0	48,100	34,900	35.0
650	1200	44,800	30,400	29.0	43,100	28,700	33.0	44,500	30,100	32.0	47,600	35,000	31.5
705	1300	42,000	27,400	35.0	41,200	25,600	38.0	42,500	27,700	35.5	44,500	31,300	36.0
760	1400	40,800	28,500	33.5	41,300	27,600	37.0	41,000	25,000	35.5	43,600	33,400	38.0
980 †	1800†	47,900	30,700	37.0	46,000	29,700	38.0	45,800	30,600	37.0	48,400	32,100	37.0
980 ‡	1800‡	46,100	28,700	35.0	45,300	25,300	32.5	46,300	29,800	34.0	48,100	29,700	30.0

* Nead and associates.

† Normalized.

‡ Normalized and box annealed at 650°C. (1200°F.).

hot-rolled strip is apparently responsible for a lower elongation in the finished sheet after annealing.

It may also be concluded that the subsequent treatments which produce the highest elongation values are annealing at 705 and 760°C. (1300 and 1400°F.) and normalizing at 980°C. (1800°F.).

After taking into consideration the tensile properties for various amounts of cold rolling and for various annealing and normalizing treatments, and the accompanying metallographic structures, and after correlating these with the Olsen-test values and Rockwell *B* hardness, Nead and his associates concluded that for the best deep-drawing properties in cold-rolled sheets, cold reduced from 0.080-in. gage bands:

1. The hot-mill finishing temperature should be 790 to 880°C. (1450 to 1615°F.).
2. As the flat finished bands had consistently better ductility than the coiled bands after cold rolling and annealing, the band should be cooled prior to coiling or should be coiled in such a manner that annealing effects are minimized.
3. The cold reductions should be 50 per cent or more.
4. The tensile strength should be between 40,000 and 46,000 lb. per sq. in., the yield ratio 60 to 70 per cent, and the Rockwell *B* hardness 40 to 50.
5. If normalizing (which produced the best structure and ductility) is not practicable, sheet annealing temperatures of at least 705°C. (1300°F.) should be used.

A brief discussion of the Olsen ductility test for deep-drawing sheet and some values for low-carbon material are given in section 256, Chapter XVII.

59. Effect of Prior Heat Treatment and Carbon Content.—In a series of comprehensive investigations, Pomp and his associates studied the effect of increasing reductions by cold working on the mechanical properties of seamless tubes, strip, and wire of varying carbon content, heat treated in various ways before cold working.

The first report, by Pomp and Albert,[(229)] on the properties of seamless tubes containing 0.10 to 0.84 per cent carbon, drawn by three methods, and after three prior heat treatments, is discussed on pages 125 to 127. The results for the three methods of drawing the tubes are given in Figs. 27 to 29. No accurate general conclusions can be drawn concerning the effect of the different heat treatments except that the elongation after the drafting seems to be independent of the heat treatment. It is

possible, of course, to come to some tentative specific conclusions by superimposing the curves for the various steels upon each other; thus, if in Fig. 29 for mandrel drawing this procedure is followed for tensile-strength curves, it might be concluded that treatment *B* results in slightly better tensile strength for steels 2 (0.28 per cent carbon) and 3 (0.41 per cent carbon) after cold working than treatment *A*, and in much better strength than treatment *C*.

Data by Pomp and Poellein(331) for cold-rolled strip of various carbon contents, subjected to a prior heat treatment which produced spheroidized cementite, are discussed on page 125 and shown in Fig. 26 reproduced from their paper. These investigators also gave curves of tensile strength and elongation for cold-drawn strip of steels 3, 6, and 9 (see composition, page 125) which had received two other prior heat treatments—annealing to form lamellar pearlite and patenting to form sorbite. The properties after cold rolling of the steels after these treatments were somewhat erratic; according to the investigators this was due to the difficulty of securing uniform structures.

For reductions by rolling of 10 to 70 per cent, steel 3, containing 0.29 per cent carbon, had slightly higher tensile strength when the original structure was lamellar pearlite, and considerably higher strength when the structure was fine pearlite (sorbite), than when the original structure consisted of spheroidized cementite. Steel 6, containing 0.65 per cent carbon, also had much higher tensile strength when originally sorbitic, but, contrary to steel 3, the strength as cold rolled was slightly higher when the original structure was spheroidized cementite than when it was lamellar pearlite.

The most recent investigation by Pomp(754) was on patented wire of the following analyses:

Steel	Composition, per cent				
	C	Si	Mn	P	S
A	0.36	0.19	0.61	0.027	0.029
B	0.44	0.16	0.62	0.014	0.037
C	0.53	0.19	0.76	0.044	0.030
D	0.62	0.17	0.69	0.015	0.041
E	0.77	0.19	0.74	0.010	0.033

These steels were hot rolled to 5.3 mm. (0.209 in.) diameter and then cold drawn to 4.6 mm. (0.181 in.) before the heat treatment and final cold drawing. The steels were quenched from 1135, 1036, 920, and 837°C. (2075, 1895, 1690, and 1540°F.) into lead which was held at 542, 460, and 406°C. (1010, 860, and

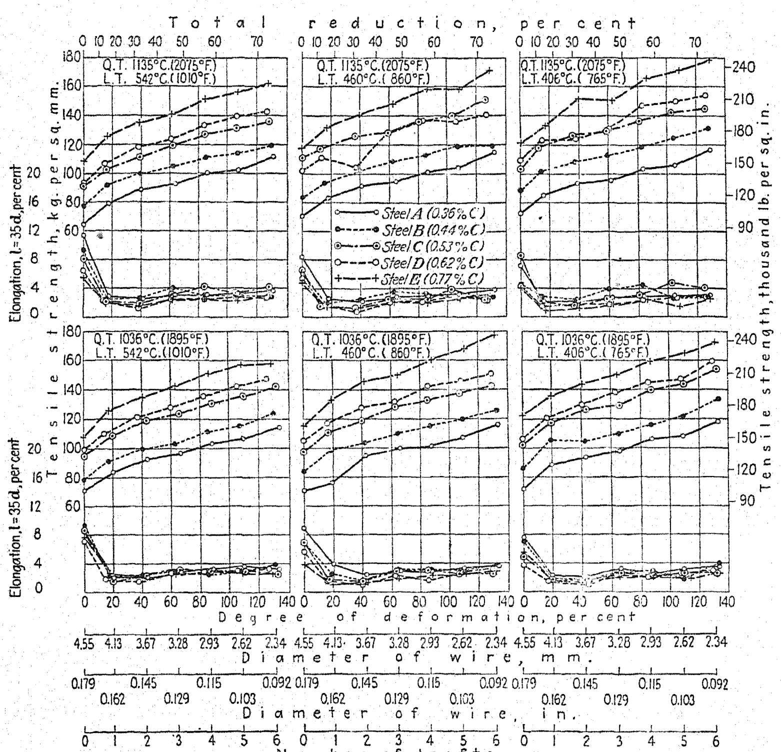

FIG. 33.—Effect of heat treatment and amount of reduction in drafting on the tensile strength and elongation of five carbon steels. Q.T. = quenching temperature; L.T. = lead temperature. (*Pomp.*[754])

765°F.). These temperatures are the averages of the actual quenching (Q.T.) and lead temperatures (L.T.) for the four lots of which the average properties are shown in Figs. 33 and 34. The former shows the tensile strength and elongation of the wires cold drawn to the degree indicated after (upper row) patenting at 1135°C. (2075°F.) into lead held at the three different temperatures and (lower row) patenting at 1036°C. (1895°F.) into lead, at the same three temperatures. The data for the same steels

patented at 920 and 837°C. (1690 and 1540°F.) respectively are given in Fig. 34.

Pomp also made reversed bend tests over a radius approximately 2.5 times the diameter of the wire. There was a general increase in the number of bends up to the fifth draft corresponding

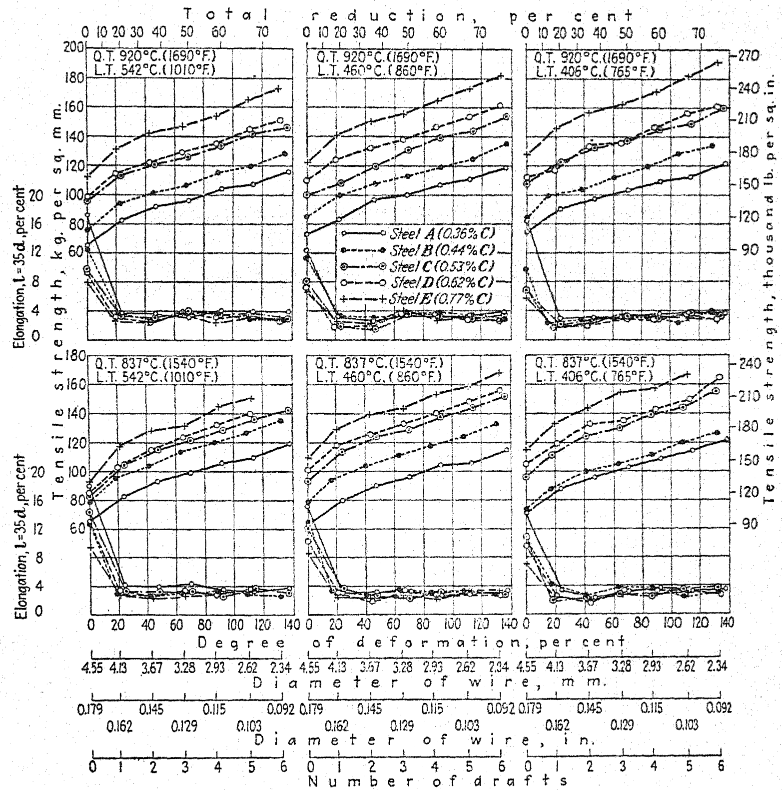

FIG. 34.—Same as Fig. 33, except for lower quenching temperatures. (*Pomp.*[754])

to a reduction in area of about 70 per cent with a decrease at the sixth draft. Torsion tests were made but the resulting curves were irregular. Using as a basis the best reversed bend value at the fifth draft, Pomp gave the values shown in Table 36 (page 144) as the best heat treatment and resulting properties.

Pomp also concluded that tensile strength is primarily controlled by the carbon content, only slightly by the quenching temperature, and is higher with the lower lead-bath tempera-

tures; that elongation decreases with increase in quenching temperature and with decrease in lead-bath temperature; and that, after the first draft, it remains approximately constant on succeeding draws.

For a high reversed bend test, Pomp found that a high quenching temperature is desirable, for a high torsion test, a medium quenching temperature, with a low lead-bath temperature in both cases.

60. Effect of Prior Heat Treatment on the Properties of Cold-worked Medium-carbon Steel.—Results which show the effect of prior heat treatment on mechanical properties, including variations in patenting practice, have been reported by Adam[(54)] for a 0.44 per cent carbon steel, by Bonzel[(792)] for a 0.40 per cent carbon steel, and by Pomp and Lindeberg[(403)] for a 0.60 per cent carbon steel. The compositions of the steels used by Adam and Bonzel were as follows:

Element	Percentage	
	Adam	Bonzel
Carbon	0.44	0.40
Manganese	0.82	0.68
Silicon	0.06	0.25
Sulphur	0.036	0.026
Phosphorus	0.031	0.020

Adam's material was a basic open-hearth steel and had been drawn two drafts from a No. 5 rod (0.212 in.) to a No. 8 wire (0.160 in.). The wire was subjected to two variations in patenting: (*a*) heating to 1000°C. (1830°F.) and air cooling and (*b*) heating to 1000°C. (1830°F.) and quenching in molten lead at 550°C. (1020°F.).

Bonzel did not give the process by which the steel was made. The first lot of wire was drawn from a 0.192-in. green rod, the second lot from a 0.188-in. rod which had been heated to 1010°C. (1850°F.) and quenched in lead at 520°C. (970°F.). The speed of the wire through the patenting furnace was 10 ft. per min. The tensile strength versus size of the wire is plotted in Fig. 35. Bonzel's data show the improvement in strength which results from patenting as compared with that resulting from drafting

an unpatented rod. Adam's data (Fig. 35) on the tensile strength of wire drawn from air-patented and lead-patented materials show that the strengths are almost the same until the seventh draft. For wire of smaller sizes than 0.06 in. lead patenting the rod resulted in higher strength. This was

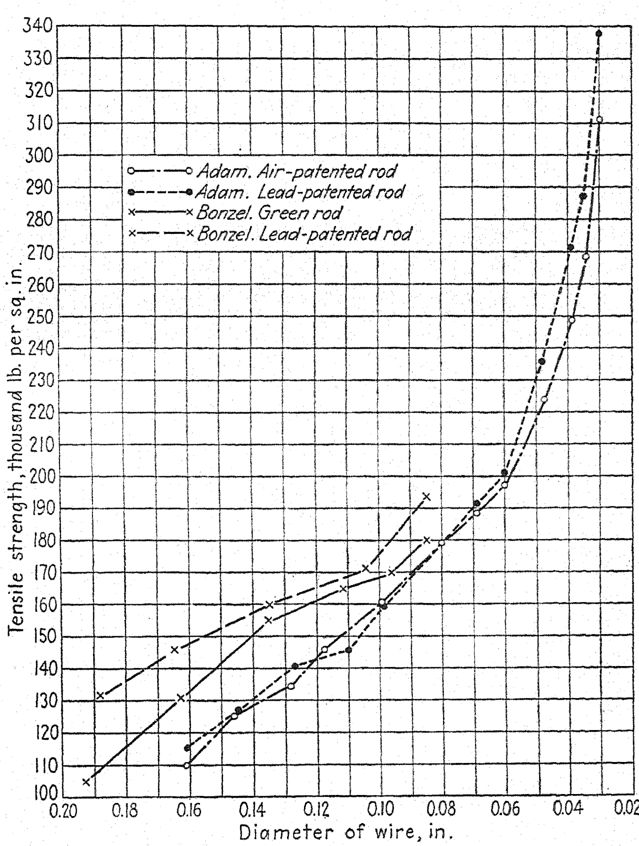

FIG. 35.—Effect of patenting on the tensile strength of medium-carbon wire. (*Adam*[54] *and Bonzel.*[792])

especially noticeable in very fine wire; the 0.03-in. wire drawn 11 drafts from lead-patented rod had a strength of 338,000 lb. per sq. in. as compared with 311,000 for wire of the same size drawn 11 drafts from air-patented rod.

Adam[54] also investigated the effect of varying the patenting temperature from 850 to 1050°C. (1560 to 1920°F.), for both air

and lead cooling, and found that the effect of temperature was slight. There were, however, indications that with air cooling the capacity of the material to withstand cold work increased with the temperature; this did not hold for lead patenting. In either practice, Adam concluded, the safe temperature range for patenting is extremely wide.

TABLE 36.—BEST HEAT TREATMENT AND THE RESULTING TENSILE STRENGTH FOR CARBON-STEEL WIRE*

Tensile-strength range, lb. per sq. in.	Steel	Carbon, per cent	Quenching temperature		Lead-bath temperature		Tensile strength, lb. per sq. in.	Reversed bends over 7.5-mm. (0.297-in.) radius
			°C.	°F.	°C.	°F.		
142,000 to 156,500	*A*	0.36	1035	1895	460	860	152,200	16
156,500 to 170,700	*A*	0.36	920	1690	405	760	159,300	15.5
	B	0.44	1035	1895	460	860	170,500	15.5
170,700 to 184,900	*B*	0.44	1135	2075	405	760	176,400	15
184,900 to 199,100	*C*	0.53	1135	2075	460	860	197,700	16
	D	0.62	1135	2075	460	860	192,500	16
199,100 to 213,400	*D*	0.62	1035	1895	405	760	209,000	16
213,400 to 227,600	*D*	0.62	920	1690	405	760	220,500	14
	E	0.77	1135	2075	460	860	224,700	14
227,600 to 241,800	*E*	0.77	1035	1895	405	760	233,300	16
241,800 to 256,000	*E*	0.77	1135	2075	405	760	243,200	16

* Pomp.(754)

Pomp and Lindeberg(403) used a steel containing 0.60 per cent carbon, 0.50 per cent manganese, 0.24 per cent silicon, and low sulphur and phosphorus. The material was in the form of a 3.45-mm. (0.136-in.) wire which had been drawn from a 5.6-mm. (0.219-in.) patented rod. Their investigation included the effect of patenting temperature and practice on the mechanical properties of the wire before final drawing and after drawing with reductions up to 90 per cent. The speed of the wire through the patenting furnace was 32 in. per min. The tests made were tensile, bend, torsion, and fatigue. The results obtained are plotted in a series of graphs. Concerning the effect of patenting

temperatures, Pomp and Lindeberg concluded that with air cooling the patenting temperature made little if any difference in the tensile strength of the cold-drawn wires. In bend strength, however, the best values resulted from the highest patenting temperature, 1070°C. (1940°F.). This was also true for torsion

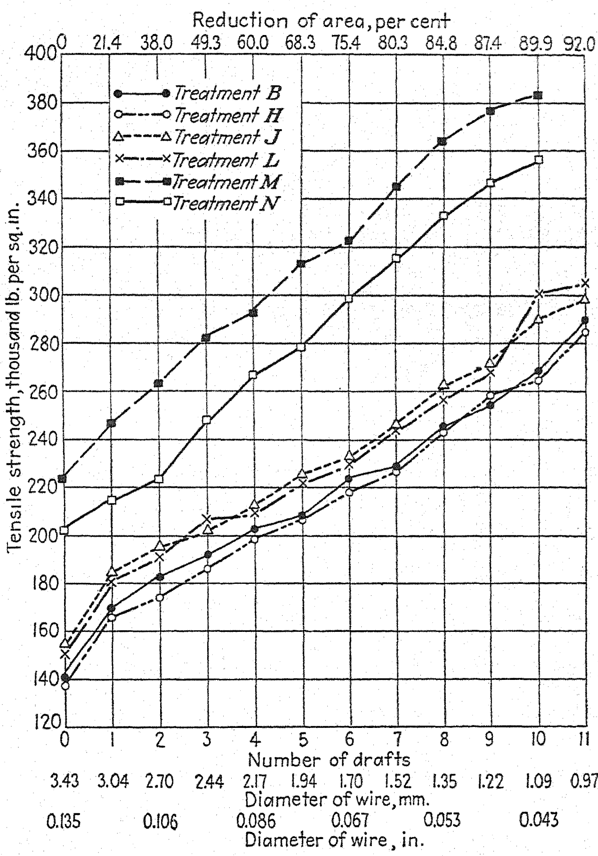

FIG. 36.—Effect of prior heat treatment on the tensile strength of cold-drawn 0.60 per cent carbon wire. For treatments see page 146. (*Pomp and Lindeberg.*[403])

results, probably because the decarburization was greater at the high temperature and the decarburized metal at the surface improved the ductility as measured by these two tests.

The patenting temperature had little if any effect on the tensile properties of the drawn wire when the original material was quenched into lead at constant temperature. Nor was tensile

strength materially affected when the original wire was quenched from a constant patenting temperature into lead of varying temperature (see Fig. 36). Bonzel,[792] on the contrary, stated that the lead bath should not be colder than 450°C. (840°F.) or hotter than 550°C. (1020°F.) as the properties are likely to be erratic, owing to the formation of martensite (with lower temperatures) or coarse pearlite (with higher temperatures) instead of the more desirable fine pearlite. In bend tests, Pomp and Lindeberg found wires drawn from lead-patented rod superior to those drawn from air-patented rod.

For wire drawn from air-patented material fatigue values were erratic and showed no clearly defined trend; for the lead-patented material, however, there was a well-defined peak at reductions in area of 60 to 80 per cent.

Pomp and Lindeberg[403] also made tests on wires drawn after quenching in oil from 980°C. (1795°F.) and tempering at 425 and 480°C. (795 and 895°F.). These wires work hardened faster (see Fig. 36) and became brittle much sooner. They were inferior to the patented wire in both bend and torsion tests.

From the large number of curves given by Pomp and Lindeberg six have been selected and plotted in Fig. 36 as representative of the tensile strength of wire cold drawn to the degree noted after the following treatments:

B. Patented at 1000°C. (1830°F.), cooled in air.

H. Patented at 930°C. (1705°F.), quenched in lead at 590°C. (1095°F.).

J. Patented at 935°C. (1715°F.), quenched in lead at 490°C. (915°F.).

L. Patented at 925°C. (1695°F.), quenched in lead at 400°C. (750°F.).

M. Quenched in oil from 980°C. (1795°F.) and tempered at 425°C. (795°F.).

N. Quenched in oil from 980°C. (1795°F.) and tempered at 480°C. (895°F.).

61. Effect of Prior Heat Treatment on Properties of Cold-worked High-carbon Steel.—A large percentage of the steels containing 0.65 per cent or more carbon (and some medium-carbon steels as well), which are cold worked, are drawn into high-strength wire to be used for cable, rope (for example for bridge construction), and springs. Practically all of this material is patented before being drawn and, if fine sizes are required, an intermediate heat treatment, usually some form of lead patenting, is given to the material during the cold-working process.

Data on the properties of wire drawn from rod which had been subjected to various prior and intermediate heat treatments have been reported by Bonzel[792] and Adam.[54] The analyses of the steels used were as follows:

Element	Percentage		
	Bonzel	Adam	
		a	*b*
Carbon	0.74	0.85	0.70
Manganese	0.68	0.30	0.54
Silicon	0.21	0.12	0.10
Sulphur	0.022	0.022	0.042
Phosphorus	0.020	0.020	0.037

Bonzel did not give the previous history of his material. One coil was untreated and was drawn six drafts from 0.200 to 0.084 in. One coil was patented (15 ft. per min.) at 950°C. (1730°F.), quenched into molten lead at 520°C. (970°F.), and drawn six drafts to 0.084 in. The third coil was annealed by heating to 850°C. (1550°F.) and cooling at 3°C. (5°F.) per min. to 400°C. (750°F.) so that the structure would consist of lamellar pearlite. This rod was drawn 10 drafts to 0.064 in. The tensile strength after each draft is plotted in Fig. 37. The improvement in strength after lead patenting is clearly evident. The strength values for the annealed rod are not strictly comparable with the values for the other two as the drafts were lighter, but they are comparable enough to indicate that lamellar pearlite is not a desirable structure for high-carbon rod which is to be drawn into high-strength wire. This was further brought out by the bend tests made by Bonzel. The wire drawn from lead-patented rod gave markedly higher bend-test values than the others.

The 0.85 per cent carbon steel used by Adam was Swedish acid open-hearth and the 0.70 per cent carbon also acid open-hearth steel but of British manufacture. Both were in the form of hot-rolled rods and were cold drawn to No. 4 (0.225-in.) wire before heat treatment prior to the regular cold working.

The specimens, whose tensile properties are plotted in Fig. 37, were treated as follows:

A. 0.85 per cent carbon steel: air patented at 1050°C. (1920°F.) and drawn seven drafts to 0.104 in.; patented again at 950°C. (1740°F.) and quenched into lead at 500°C. (930°F.); then drawn nine drafts to 0.023 in.

B. 0.85 per cent carbon steel: same treatment as under *A* except that the temperature of the molten lead was 600°C. (1110°F.) and the wire was drawn eight drafts to 0.027 in.

C. 0.70 per cent carbon steel: same treatment as under *A* except that the wire was drawn eight drafts after the lead patenting to 0.027 in.

As is evident from Fig. 37, the 0.85 per cent carbon steel (treatment *A*) had a tensile strength for all reductions in area

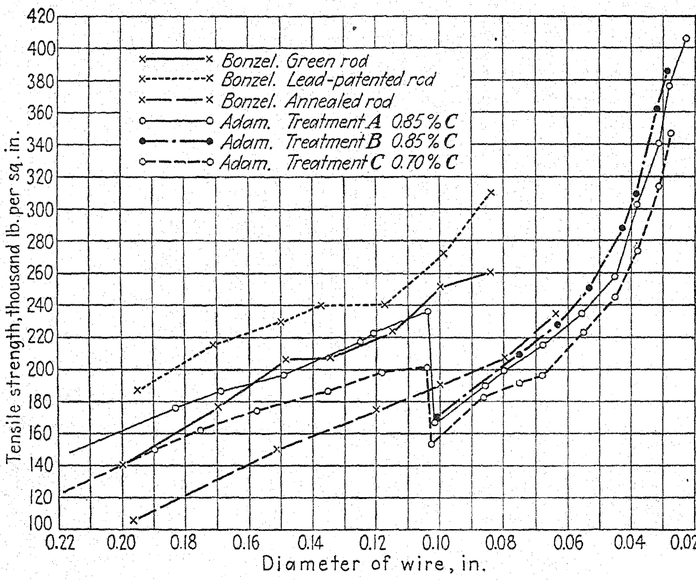

FIG. 37.—Effect of prior and intermediate heat treatments on the tensile strength of cold-drawn high-carbon wire. (*Adam*[54] *and Bonzel.*[792])

that was 10,000 to 25,000 lb. per sq. in. higher than the strength of the 0.70 per cent carbon steel (treatment *C*). Quenching in lead maintained at 600°C. resulted in slightly higher tensile-strength values (treatment *B*) than quenching the same steel in lead at 500°C. (treatment *A*). The increase, however, was not great.

From the complete series of tests made by Adam he concluded that, as in the case of the medium-carbon steel (see page 144), the

permissible range of patenting temperatures is extremely wide, and that "the improvement obtained by lead patenting is even more striking in these [high-carbon] steels than in the medium-carbon steel." According to Adam "the fine structure produced by quenching in oil and tempering is quite unsatisfactory for withstanding cold flow" and "ordinary annealing is not suitable as a heat-treatment operation prior to cold working this class of steel."

This unsuitability of prior annealing is confirmed by the results of Swinden and Bolsover.[239] Steel *D* containing 0.70 per cent carbon (see composition, page 123), in the form of 0.300-in. sheet, was normalized by air cooling from 830°C. (1525°F.) and annealed 4 hr. at 750°C. (1380°F.) which was followed by cooling for 10 hr. The normalized and the annealed sheets were cold rolled with the same reductions in thickness, given in Table 33 (page 124), and the mechanical properties were determined. Properties of the normalized and cold-rolled strip checked closely with the properties of the hot-rolled and cold-rolled strip; this might have been expected as steel of 0.70 per cent carbon is probably in the normalized condition after air cooling from the finishing temperature of hot rolling. The properties of the annealed and cold-rolled strip were much inferior to those of the corresponding specimens cold rolled from the hot-rolled and the normalized sheet. Some typical values are as follows:

Prior condition	Reduction by cold rolling, per cent	Tensile strength, lb. per sq. in.	Elongation, in $4\sqrt{A}$, per cent
Hot rolled	None	128,900	22.1
Normalized	None	133,900	19.4
Annealed	None	92,900	26.8
Hot rolled	26	160,300	10.7
Normalized	26	161,000	12.1
Annealed	26	137,000	7.5
Hot rolled	67	198,300	8.2
Normalized	67	196,900	9.3
Annealed	67	174,900	5.5

62. Mechanical Properties of Spring Wire.—There are three principal grades of carbon-steel wire used for springs,* known commonly as hard-drawn wire, oil-tempered wire, and music wire. Eakin[691] gave the following details:

Hard-drawn spring wire is usually drawn from basic open-hearth steel containing 0.50 to 0.70 per cent carbon, 0.70 to 1.10 per cent manganese, 0.10 to 0.20 per cent silicon, and less than 0.04 per cent sulphur and phosphorus. The rods are drawn, with or without prior patenting, to finished size if large, or with an intermediate annealing or lead patenting for small sizes.

Oil-tempered spring wire is drawn from basic open-hearth steel of about the same analysis as the hard-drawn wire. As oil-tempered wire is usually fairly large in size, the practice consists of drawing the rods, after annealing or patenting, to finished size and then heat treating by quenching in oil and tempering in oil or lead. The wire is not cold worked further after this heat treatment.

Music wire is the highest grade of wire made. The rods are rolled from carefully selected acid open-hearth or electric-furnace steel containing 0.70 to 1.00 per cent carbon, 0.25 to 0.50 per cent manganese, 0.10 to 0.20 per cent silicon, and less than 0.025 per cent sulphur and phosphorus. The rods are given a prior patenting, and are then cold drawn, lead patented, and drawn again to finished sizes, and may have in the smaller sizes tensile strengths exceeding 400,000 lb. per sq. in.

There is also a fourth variety, known as heat-treated high-carbon springs. These are made from open-hearth or electric steel wire containing 0.80 to 1.00 per cent carbon and 0.35 to 0.45 per cent manganese, drawn to size, with or without prior annealing or patenting, and annealed. The springs are then formed and are heat treated (spring tempered) after forming. The tensile properties of this material are about the same as, or a little higher (because of the higher carbon content) than, those of oil-tempered wire.

Representative tensile strengths plotted against the size of the wire for the three grades just described are given in Fig. 38. The curves marked music wire, oil-tempered wire, and hard-drawn wire are from Eakin, the curve marked Premier from

* A number of alloy steels and a few non-ferrous alloys are also used for springs.

unpublished data* on the properties of a high-grade acid open-hearth spring wire containing about 0.65 per cent carbon and 1.00 per cent manganese, made by a prominent American wire manufacturer.

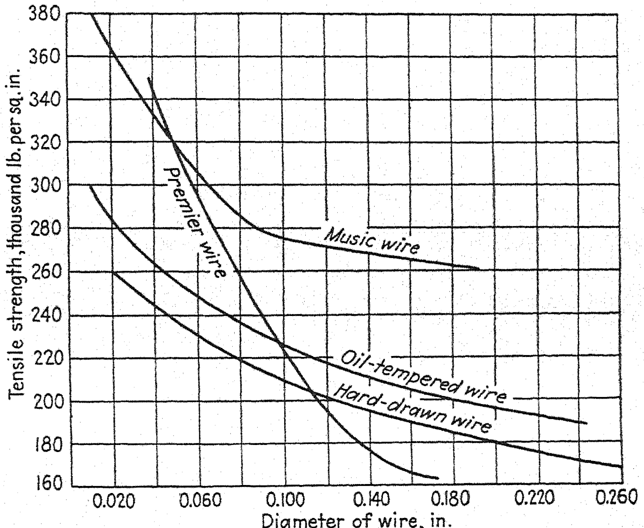

FIG. 38.—Tensile strength for various sizes of spring wire. (*Eakin*[691] *and Sisco.*)

D. EFFECT OF AGING AND SUBSEQUENT HEAT TREATMENT

Cold-worked products may, after the final cold drawing or rolling, receive any of a variety of heat treatments ranging from reheating to slightly above room temperature, or annealing at the recrystallization temperature—400°C. (750°F.) or above, depending upon composition, amount of cold work, and other variables—to annealing above the critical range; or they may be quenched and tempered, depending upon the mechanical properties or structural condition desired in the finished product. In addition, cold-worked steel may age harden at room temperature if given enough time.

In general, thermal treatments which produce almost complete recrystallization—full annealing or quenching and tempering—affect the properties of cold-worked steel in much the same way as they affect those of hot-worked material. Typical tensile

* Sisco.

properties of annealed low-carbon Bessemer and basic wires are given in Tables 31 and 32 (pages 117 and 118), and of quenched and tempered (oil-tempered) high-carbon wire in Fig. 38. Except for the brief discussion of annealing given later in section 66, the effect of thermal treatments which produce complete recrystallization is given no further attention in the present chapter; a complete discussion of these effects is contained in the next chapter.

63. The Effect of Aging at Normal Temperature on Properties. The mechanism of aging has been discussed in detail in Volume I of this monograph (Chapter XII, pages 384 to 396). As noted there, the spontaneous increase in hardness at room temperature with time is primarily a precipitation effect which occurs frequently in low-carbon steels after quenching (quench aging) or after straining (strain aging). Strain hardening also occurs in pure single-phase metals. Quench aging is not important in connection with a discussion of the properties of cold-worked steels; strain aging is important as its effects, the changes in mechanical properties caused by precipitation of carbide or other particles, are induced by the cold working. The same change in properties caused by aging for a relatively long time at normal temperature may take place in a few minutes (for small sections) if the material is heated slightly.

The effect of strain aging on the properties of high-purity iron and very low carbon steel (about 0.02 per cent carbon) has been discussed in a previous monograph[794] and is not repeated here. In Volume I of this monograph (Chapter XII, page 394) Epstein quoted data which indicate that the increase in hardness after strain aging is less than that caused by quench aging, but the embrittlement due to strain aging may be greater. Thus, cold working a low-carbon "aging" steel reduced the impact value to less than 10 ft-lb. as compared with a value of about 40 ft-lb. for a "non-aging" steel similarly cold worked.

A series of tests on the effect of aging on mechanical properties was reported by Adam.[145] The steel used contained 0.11 per cent carbon, 0.53 per cent manganese, 0.05 per cent silicon, 0.037 per cent sulphur, and 0.051 per cent phosphorus, and was drawn into wire large enough so that the impact resistance could be determined. The diameter of the rod, the diameter of the wire, reduction in area in drawing, and speed of drawing for five

of the eight series reported by Adam are given in Table 37. The reduction was carried out in one draft. Specimens were tested as drawn, no machining was done other than milling the notches for the Izod specimens.

Some of Adam's results are given in Table 38. Aging increased the tensile strength of this particular steel 10 to 15 per cent (with the exception of series *H* where the increase was only 6 per cent) and the yield strength between 10 and 20 per cent. Proportional limit was increased 250 to 300 per cent. The ductility was reduced; this is most marked in the impact values for all but series *E* (not reproduced) and *F*; but the values for elongation and reduction of area were also lowered appreciably. Adam also gave values for modulus of elasticity. Aging increased the modulus of every series from 3 to 9 per cent, corresponding to an actual increase of 780,000 to 2,350,000 lb. per sq. in.

TABLE 37.—DRAFTING PRACTICE FOR WIRE CONTAINING 0.11 PER CENT CARBON, THE PROPERTIES OF WHICH ARE GIVEN IN TABLE 38*

Series	Original diameter, in.	Reduced diameter, in.	Reduction of area in drafting, per cent	Speed of drawing, in. per min.
B	0.4704	0.4490	8.7	3.163
D	0.5293	0.4490	28.02	2.988
F	0.5897	0.4492	41.95	2.998
H	0.6548	0.4490	52.98	2.947
I	0.6861	0.4476	57.45	2.912

* Brown, according to Adam.(145)

In a study of the effect of aging on low-carbon steel at atmospheric temperature, Pfeil(278) found that the maximum aging effect was produced (with sections from 0.06 to 0.50 in. diameter) in about one month. This is not in agreement with the results reported by Adam(145) (Table 38) which show that in most of the series there was an appreciable increase in strength upon aging from 28 to 200 days. Moreover, Pfeil's results do not agree with the results of Köster, von Köckritz, and Schulz,(539) discussed on page 157. These investigators found that in the sample aged at normal temperature there was an increase in tensile strength up to the third month.

Pfeil determined tensile strength only and found that the increase resulting from aging wire which had been reduced 98 per

TABLE 38.—EFFECT OF AGING ON THE MECHANICAL PROPERTIES OF LOW-CARBON WIRE DRAWN AS INDICATED IN TABLE 37*

Series	Interval between drawing and testing, days	Tensile strength, lb. per sq. in.	Yield strength,† lb. per sq. in.	Proportional limit, lb. per sq. in.	Elongation in $4\sqrt{A}$, per cent	Reduction of area, per cent	Izod-impact value, ft-lb.
A	‡	55,800	32,400	23,500	43.83	67.50	99.34
B	1/24	59,800	51,700	9,700	36.80	70.84	47.5
	1	61,600	54,000	21,100	26.44	68.31	43.3
	8	64,600	59,100	27,700	28.30	67.48	32.1
	28	67,000	62,000	33,100	25.15	67.56	19.8
	223	68,800	61,900	32,900	25.15	65.33	16.3
D	1/24	74,400	65,700	11,600	22.02	62.23	9.2
	2	76,400	70,800	13,200	22.02	60.68	5.2
	7	79,500	74,600	23,500	18.87	62.03	4.8
	24	84,900	79,100	26,400	15.72	56.98	4.3
	216	83,900	79,500	29,300	18.87	58.03	4.9
F	1/24	84,300	77,800	14,400	18.87	55.63	7.32
	1	86,600	80,800	19,900	18.24	62.22	6.78
	7	91,700	86,800	22,800	15.10	56.59	4.07
	26	91,800	88,500	31,400	13.84	50.46	5.00
	188	95,800	92,300	33,800	13.84	50.46	7.47
H	1/24	92,700	86,800	15,300	16.35	50.63	26.9
	1	92,500	87,000	19,700	15.72	51.28	13.1
	7	96,700	94,800	26,600	12.58	47.86	7.5
	28	97,400	94,800	33,600	12.89	48.37	8.69
	190	99,000	96,100	33,800	13.20	48.99	10.3
I	1/24	90,000	87,200	15,100	13.55	49.20	32.6
	1	98,100	92,500	17,700	14.29	49.80	29.12
	7	97,900	96,700	24,600	12.58	51.40	13.87
	26	....	....	......	.....	.....	18.82
	191	101,900	100,600	32,700	11.73	47.26	

* Brown, according to Adam.(145)
† Reported as yield point.
‡ Annealed bar.

cent by drafting varied from 5200 lb. per sq. in. for wire containing 0.11 per cent carbon to 7800 lb. per sq. in. for wire containing 0.014 per cent carbon, or 4 to 10 per cent. The increase was apparently independent of the amount of reduction in drafting.

The results obtained by Adam and by Pfeil on the effect of varying reductions in drafting were not confirmed by a recent report by Hudson.[803] In connection with work on the corrosion of hard-drawn wire, it was found advisable to determine the effect of aging, if any, on mechanical properties. Samples of wire containing 0.22 per cent carbon, 0.60 per cent manganese, and 0.04 per cent silicon were sent to eight laboratories for tests as received and after aging for one year. It was found that the actual breaking load of wire approximately 0.25 in. in diameter, which had undergone a 24 per cent reduction of area in drafting, had increased 3.7 per cent after aging one year, and the elongation value had fallen 14 per cent. In the case of smaller wires (about 0.135 and 0.065 in. in diameter), which had received 78 and 75 per cent reduction respectively, no definite change in the mechanical properties could be detected. Hudson considered "that the change in properties of hard-drawn wire upon aging is determined to some extent by the degree of previous cold working."

From these results it may be concluded that, if low-carbon steel is of the quality known, for lack of a better name, as "aging steel," cold-worked sections will age at atmospheric temperature which results in somewhat higher tensile- and yield-strength values and lower elongation, reduction of area, and impact resistance. Under the same conditions, the proportional limit may be doubled or even tripled.

Data on the relation between amount of reduction in cold working and the effect of aging on the properties are so erratic that no conclusions can be formulated. This is discussed in section 66.

64. Effect of Low-temperature Heating on Properties of Wire. While the general effects of cold working are to increase hardness and strength and decrease plasticity, various properties are affected in various ways. Thus, according to Jeffries and Archer,[125] "the elastic limit is sometimes lowered to zero by cold working but may be restored by aging or heating at low temperatures to a value which may be much higher than its original value. . . . The yield strength is probably affected more by cold work than any other property of the hardness group." These phenomena are clearly shown in Table 38; cold working reduces the proportional limit to a low value, but aging increases it to a value much higher than the original

(annealed) value; cold working increases the yield strength until it is almost equal to the tensile strength (compare the unaged yield-strength values for series *B* and series *H* and *I*).

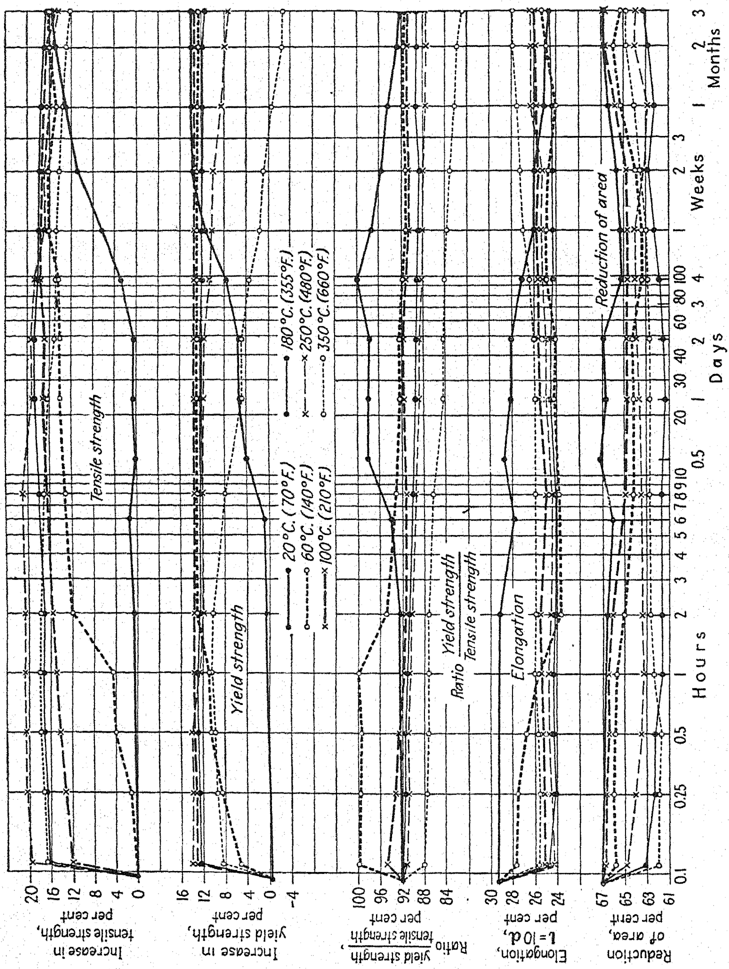

FIG. 39.—Effect of reheating temperature and time on the properties of cold-worked steel containing 0.10 per cent carbon and 0.41 per cent manganese. (*Köster, von Köckritz, and Schulz.*[539])

Reheating to temperatures between 100°C. (210°F.) and the recrystallization temperature speeds up these changes so that they take place in a few minutes instead of days or even months. This is shown by the investigations of Köster, von Köckritz, and Schulz,[539] Greaves,[120] and Rees.[107]

The first-named used a steel with 0.10 per cent carbon, 0.41 per cent manganese, 0.01 per cent silicon, 0.033 per cent sulphur, 0.026 per cent phosphorus, 0.10 per cent copper, and 0.006 per cent nitrogen. Specimens 10 mm. (0.394 in.) in diameter by 100 mm. (3.94 in.) gage length were cold worked by stretching until they elongated 18 per cent. They were then aged at various temperatures for various times. Results, plotted on a logarithmic scale, are given in Fig. 39. These show that, for the steel used, for the method of cold working, for the size of specimen and times of heating, the tensile strength of the specimens held

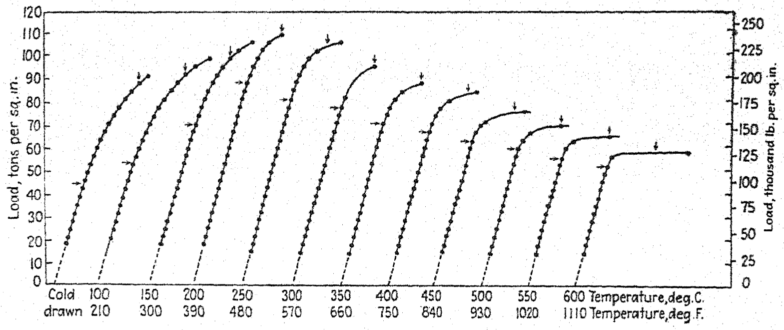

FIG. 40.—Stress-strain curves for a cold-drawn 0.74 per cent carbon steel, reheated 1 hr. at the indicated temperatures. (*Greaves.*[120])

at normal temperature did not increase materially until after 3 or 4 days of aging. Yield strength was affected sooner, increasing after about 6 hr. Aging at 60°C. (140°F.) sped up the changes to about 1 hr. for the tensile, and a few minutes for the yield strength. Heating to 100°C. (210°F.) or above produced aging in a very few minutes.

The greatest increase in tensile and yield strengths resulted from heating at 250°C. (480°F.), but holding at this temperature caused a progressive lowering of both tensile and yield strengths after 6 to 12 hr. This is even more marked in the specimens heated at 350°C. (660°F.), especially in the yield strength which decreased after heating for 1 to 2 hr. Values for reduction of area were lowered almost immediately by reheating but were affected very little more by increasing the heating time. For example, after 10 min. heating at 180 and 250°C. (355 and 480°F.) the values for reduction of area were 62.5 and 64.3 per cent respectively. After 3 months at these temperatures the values were 63 and 64 per cent.

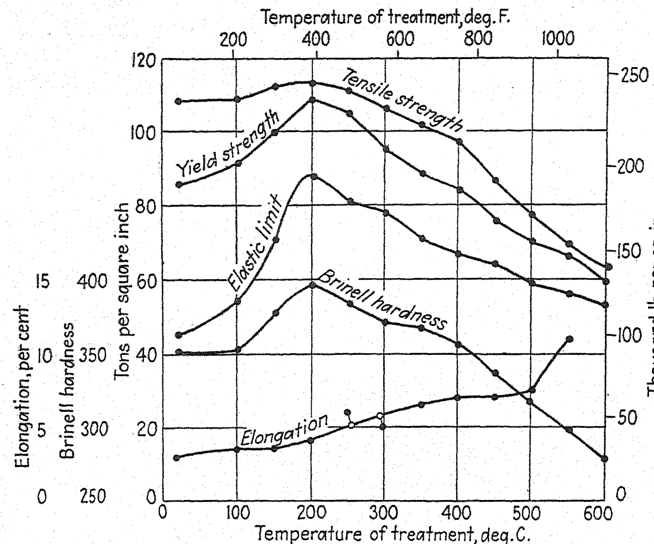

FIG. 41.—Tensile properties of a cold-drawn 0.74 per cent carbon steel, reheated 1 hr. at the indicated temperatures. (*Greaves.*[120])

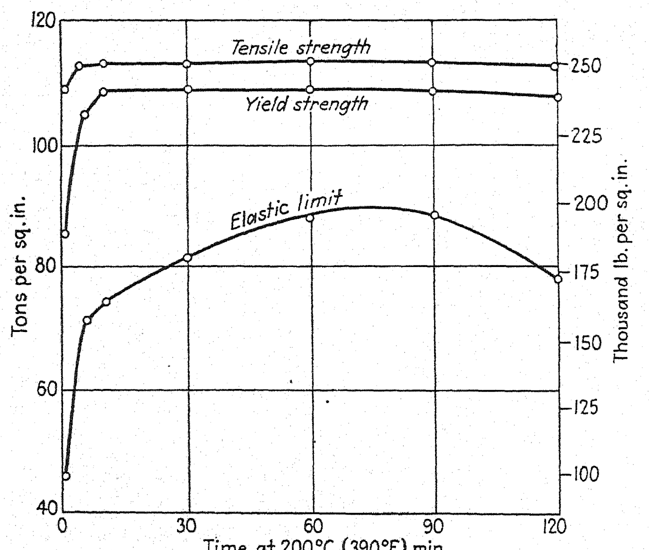

FIG. 42.—Effect of time of reheating at 200°C. (390°F.) on the tensile properties of 0.74 per cent carbon steel. (*Greaves.*[120])

Greaves[120] studied the effect of reheating on the properties of a cold-drawn high-carbon gun wire containing 0.74 per cent carbon, 0.74 per cent manganese, and 0.14 per cent silicon. His stress-strain curves are reproduced in Fig. 40; the horizontal arrows show the elastic limit (for 0.005 per cent permanent set) and the vertical arrows the yield strength (representing 1 per cent total deformation). The effect of reheating for 1 hr. at the temperatures shown is clearly indicated. The maximum tensile and yield strengths for this steel resulted from a reheating temperature of 200°C. (390°F.). Values for the effect of reheating for 1 hr. on tensile properties and hardness are plotted in Fig. 41. Greaves also determined the effect of time at 200°C. (390°F.) on the tensile strength, yield strength, and elastic limit. These values are plotted in Fig. 42.

TABLE 39.—PROPERTIES OF COLD-DRAWN STEELS ANNEALED AT LOW TEMPERATURES FOR 1 HR.*

Treatment	Tensile strength, lb. per sq. in.	Porportional limit, lb. per sq. in.	Yield strength,† lb. per sq. in.	Elongation in 2 in., per cent	Reduction of area, per cent
Steel containing 0.04 per cent carbon					
Before drawing	56,400	29,100	38,100	36.5	69
After drawing	58,800	44,800	72,600	17.0	54
Annealed at 100°C. (210°F.)	80,600	49,300	71,600	21.0	59
Annealed at 200°C. (390°F.)	79,800	47,000	71,600	20.0	57
Annealed at 300°C. (570°F.)	78,800	56,000	71,600	21.0	55
Annealed at 450°C. (840°F.)	79,800	51,500	61,800	26.0	66
Annealed at 550°C. (1020°F.)	68,100	49,300	60,000	28.0	66
Annealed at 600°C. (1110°F.)	61,800	24,600	50,500	30.0	69
Annealed at 650°C. (1200°F.)	52,900	15,700	28,200	42.0	73
Steel containing 0.85 per cent carbon					
Before drawing	138,900	56,000	69,400	15.0	22
After drawing	159,000	49,300	125,500	6.0	18
Annealed at 100°C. (210°F.)	161,300	53,500	136,600	5.0	17
Annealed at 200°C. (390°F.)	168,000	107,500	150,000	5.0	14
Annealed at 300°C. (570°F.)	168,000	109,800	145,500	5.0	14
Annealed at 450°C. (840°F.)	156,800	98,600	116,500	8.5	17
Annealed at 650°C. (1200°F.)	130,000	71,700	80,600	14.0	29
Annealed at 700°C. (1290°F.)	116,500	40,300	56,000	23.5	38

* Rees.[107]

† Stress for a permanent change in dimension of 0.2 per cent.

The effect of reheating two steels, containing 0.04 and 0.85 per cent carbon respectively, for 1 hr. at various temperatures was investigated by Rees.[107] His results are given in Table 39. Both steels received a relatively small amount of cold working; the low-carbon material was reduced 14.8 per cent and the high-carbon steel only 8.1 per cent. The reheating treatment did not affect greatly the proportional limit of the low-carbon steel, but in the high-carbon material this property was almost doubled by heating to 200 to 300°C. (390 to 570°F.).

Although galvanizing is, strictly speaking, not a reheating process, its effect upon the properties of cold-worked material should be mentioned briefly. In passing rapidly through a bath of molten zinc, the surface, and probably the material near the surface of the steel, becomes heated to approximately 470°C. (880°F.). The tensile strength is usually decreased and the elongation increased.* Adam[145] gave the following average percentage changes of properties in galvanizing wires containing 0.12, 0.60, and 0.80 per cent carbon, after drawing 2, 3, 4, 5, and 6 drafts:

Carbon, per cent	Percentage change in			
	Tensile strength	Elongation	Torsion	Bend
0.12	−12	+200	− 8	0
0.60	− 7.6	+ 85	−31	−43
0.80	−10	+ 7.1	−63	−57

The embrittling effect of galvanizing is evident in the lowered number of torsions and in the lowered bend values in the high-carbon wire. The embrittlement caused by galvanizing is discussed in Volume I, section 137, of this monograph as a possible factor in the failure of high-carbon bridge wire.

65. Effect of Low-temperature Heating on Properties of Cold-rolled Strip.—Swinden and Bolsover[239] investigated the effect of heating for 1 hr. at 100 to 700°C. (210 to 1290°F.) on the mechanical properties of strip steel which had been reduced

* This is true for cold-drawn material, but on annealed sheets the effect of galvanizing is just the reverse; the tensile strength is increased and the elongation decreased by galvanizing.

33 and 67 per cent in thickness (from 0.30 to 0.20 and 0.10 in.) by cold rolling. The complete analyses of the steels used are given on page 123. Data on the tensile strength and elongation are given in Table 40. Minimum proof stress (0.1 per cent permanent elongation), shear strength, and Brinell hardness were also determined.

TABLE 40.—EFFECT OF HEATING ON THE PROPERTIES OF COLD-ROLLED STRIP*

Reheated 1 hr. at		Steel *A* (0.10 % carbon)		Steel *B* (0.34 % carbon)		Steel *C* (0.54 % carbon)		Steel *D* (0.70 % carbon)	
°C.	°F.	Tensile strength, lb. per sq. in.	Elongation in $4\sqrt{A}$, per cent	Tensile strength, lb. per sq. in.	Elongation in $4\sqrt{A}$, per cent	Tensile strength, lb. per sq. in.	Elongation in $4\sqrt{A}$, per cent	Tensile strength, lb. per sq. in.	Elongation in $4\sqrt{A}$, per cent
Reduced 33 per cent in thickness									
Cold rolled		84,500	15.0	126,200	10.0	146,000	8.4	163,300	11.4
100	210	84,500	14.0	125,700	12.7	146,700	9.3	168,900	10.7
200	390	85,000	16.7	129,700	12.0	151,400	9.3	175,800	11.3
300	570	83,800	17.3	130,800	14.0	152,100	10.7	174,300	12.0
400	750	79,300	22.0	123,900	16.0	144,700	14.0	166,500	12.7
500	930	73,400	26.7	115,100	20.7	133,200	14.7	155,300	16.7
600	1110	51,900	50.0	89,400	32.0	112,900	23.3	135,500	20.7
700	1290	51,700	48.0	78,000	38.1	91,800	32.0	113,100	26.0
Reduced 67 per cent in thickness									
Cold rolled		103,000	8.7	154,100	8.2	181,800	6.4	198,400	8.2
100	210	103,200	11.1	153,400	7.3	182,200	8.2	204,600	9.1
200	390	101,700	10.9	154,400	8.2	187,500	9.1	216,400	10.0
300	570	100,600	10.0	155,700	11.8	187,000	10.0	218,000	10.9
400	750	94,500	20.9	143,600	15.5	173,800	11.8	197,100	11.8
500	930	84,000	21.8	121,200	18.2	144,700	16.4	165,700	13.6
600	1110	55,500	56.4	89,400	36.4	110,700	25.4	135,100	20.0
700	1290	51,300	54.5	75,500	40.0	89,400	37.3	108,600	29.1

* Swinden and Bolsover.(239)

Swinden and Bolsover concluded from their results that the tensile strength increases with heating temperature, in all specimens except the 0.10 per cent carbon strip, until approximately 300°C. (570°F.) is reached and decreases with higher temperatures. Proof stress increases and decreases at approximately the same rate as the tensile strength.

It will be noted from Table 40 that the elongation remained the same or decreased slightly when the material was heated to 300°C. (570°F.) or below. With higher heating temperatures the elongation increased slowly until the recrystallization temperature was reached, 600°C. (1110°F.) or above, when it increased rapidly to a percentage characteristic of annealed material. In general it was found that the percentage reduction of area of the test piece followed the same trend as the elongation.

Brinell-hardness values were somewhat irregular; in general, they increased with tensile strength but at a somewhat greater rate, thus reducing for the higher temperatures the ratio between these two values. For example: steel *A* had a hardness over tensile-strength ratio of 0.227 and 0.220 as cold rolled to 33 and 67 per cent reduction of thickness, and of 0.208 and 0.194 after heating 1 hr. at 700°C. (1290°F.). There were similar lowered ratios for the other steels. Shear stress followed the same trend as tensile strength, but neither the increase for temperatures up to 300°C. (570°F.) nor the decrease for higher temperatures was so great as in tensile strength and proof stress.

Swinden and Bolsover also determined the effect of time of holding at 700°C. (1290°F.) on the properties of steel *D* (0.70 per cent carbon) after a 76 per cent reduction in thickness in cold rolling. The tensile strength decreased from 125,400 lb. per sq. in. for 30 min. at 700°C. to 86,500 lb. per sq. in. for 6 hr. Shear stress also decreased, but percentage elongation and reduction of area were but slightly affected.

66. Davenport and Bain's Conclusions Regarding Strain Aging.—One of the noteworthy characteristics of the data on aging at room temperature and on the effect of slightly elevated temperatures on the mechanical properties of cold-worked steel is the inconsistency of many of the results. For example, Pfeil, Adam, Hudson, and others (see pages 152 to 155) do not agree upon the time when the maximum aging effect (at room temperature) is apparent, or on the relation between age hardening and the amount of reduction in drafting.

In a recent report[795] of an elaborate study of aging, Davenport and Bain concluded that hardening due to strain aging was caused by the fact that ferrite, apparently supersaturated with oxygen, rejects an iron-oxygen compound in the slip bands. These investigators also showed that the effects of strain aging

can be lessened by a pre-aging treatment [for example, prior heating to 100°C. (210°F.) or higher, depending upon the material]. If Davenport and Bain's conclusions are accepted, the reason for the erratic results on age hardening by the various investigators becomes clearer. Obviously, the amount of age hardening which takes place at room (or higher) temperature depends on two widely varying factors: the amount of oxygen in the steel, and the treatment prior to cold working. The last variable, especially, may be responsible for erratic results, even in the same investigation. Davenport and Bain showed that pre-aging at one temperature prevented any increase in hardening upon subsequent aging, but pre-aging at a higher temperature did not prevent subsequent age hardening.

Obviously, to determine the effect of such a variable as the amount of cold reduction on subsequent age hardening, all of the specimens must be from the same heat and must contain, presumably, the same amount of oxygen. Moreover, the thermal treatment prior to cold working must be carefully controlled so that subsequent age hardening will not be inhibited to a greater extent in one specimen, or one series of specimens, than in another. If such careful control of these variables is necessary to secure consistent results in a single investigation, it is obvious that accurate conclusions on the effect of age hardening on tensile properties from two or more different investigations are exceedingly difficult to obtain.

67. Softening by Annealing.—The annealing temperatures and the time at a given temperature required to remove completely the hardness resulting from cold work depend upon the composition of the steel and the amount of cold work; the temperature decreases as the amount of reduction by rolling or drafting increases. The following shows temperatures used commercially in the bright annealing of low-carbon hard-drawn wire:

Size of wire, in	Annealing temperature	
	°C.	°F.
0.080 or larger	620 to 650	1150 to 1200
0.062 to 0.018	480 to 510	900 to 950
0.016 or smaller	400 to 425	750 to 800

Low-carbon strip and sheet are annealed at 600 to 760°C. (1110 to 1400°F.); the higher ranges are used for black annealing and the lower temperatures for bright (scale-free) annealing.

A large proportion of the tonnage of cold-drawn wire and cold-rolled strip is annealed before being used. The properties of such material depend chiefly upon the carbon content and, as already stated, do not differ greatly from the properties of hot-rolled and annealed carbon steels discussed in the next chapter; therefore, they have received no more than a casual mention here. Typical properties of annealed low-carbon wire are given in Tables 31 and 32 (pages 117 and 118); those of a 0.04 per cent and a 0.85 per cent carbon steel, annealed at 600 and 650°C. (1110 and 1200°F.) in Table 39 (page 159); those of a 0.74 per cent carbon-steel wire in Fig. 41; and those of strip of varying carbon content in Table 40 and on page 162. Typical properties of commercial cold-rolled and annealed automobile-body sheet are given in Table 35 (page 137). Although the tensile properties alone do not determine the suitability of annealed low-carbon sheet for deep drawing, the best range of values is, as stated on page 138, 40,000 to 46,000 lb. per sq. in tensile strength, 60 to 70 per cent yield ratio, and 40 to 50 Rockwell *B* hardness.

Data on properties showing the correlation between the annealing temperature and the amount of reduction in cold-rolled high-carbon strip, containing 0.65 per cent carbon, 0.28 per cent manganese, and 0.04 per cent silicon, are given in Fig. 43 from Pomp and Poellein.[331] The specimens were of the thickness given in Fig. 43 and had a 50-mm. (1.97-in.) gage length; they were annealed for 4.5 hr. at the temperatures shown. Considerable softening was evident in the specimens heated to 550 and 600°C. (1020 and 1110°F.), and complete softening occurred when the temperature was 650°C. (1200°F.) and above.

If steels of a very low carbon content are cold rolled with a small reduction in cross-section and then reheated to within a certain temperature range, very large ferrite grains are formed, and the steel, or ingot iron, has properties characteristic of a coarse-grained material. According to Körber,[89] annealing at temperatures between 650°C. (1200°F.) and the A_3 temperature produces very coarse grains in "critically strained" ferrite. Kinský[627] recently outlined the conditions responsible for the formation of coarse columnar crystals in sheet steel and pointed

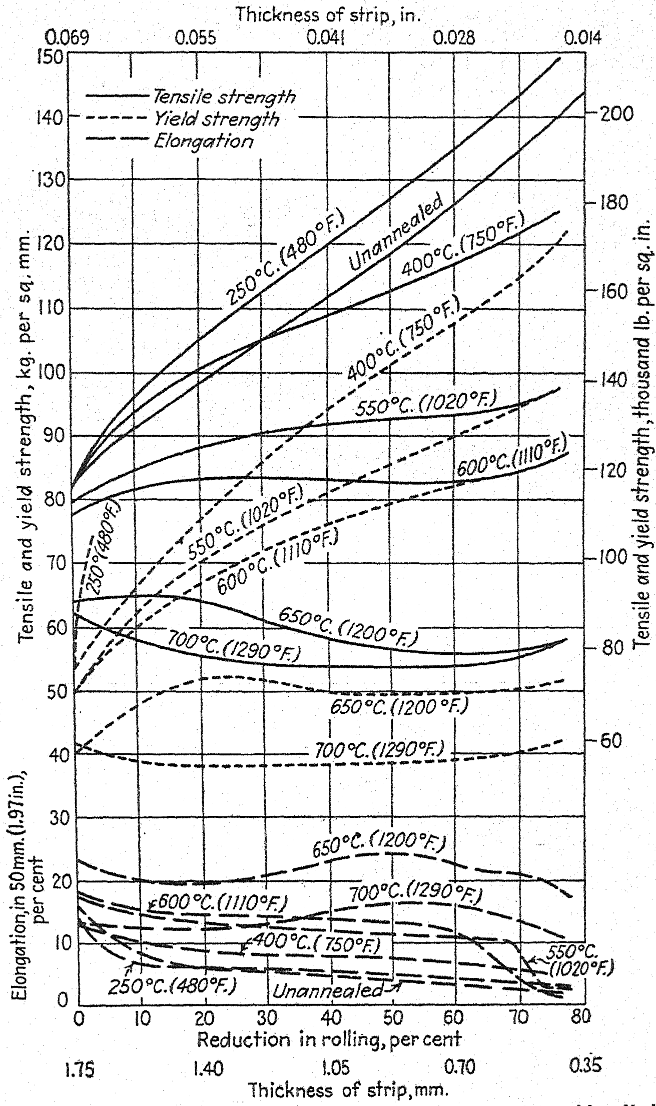

FIG. 43.—Tensile properties of 0.65 per cent carbon strip, cold rolled with the reductions shown and annealed 4.5 hr. at the indicated temperatures. (*Pomp and Poellein.*[331])

out that a decarburized surface was a prerequisite for the formation of such crystals. Recrystallization, grain growth, and the effect of annealing after cold working in high-purity iron and iron-carbon alloys containing 0.02 or 0.03 per cent or less carbon have been discussed in detail in a previous monograph of this series.[794]

Nead, Mahlie, and Dittrich,* in their investigation of the effect of rolling and annealing on the properties of 0.05 per cent carbon continuous-mill strip, found critical straining after annealing at 650°C. (1200°F.) in bands, flat rolled with a finishing temperature of 880°C. (1615°F.) after cold rolling with reductions of 5 to 15 per cent, and after annealing at 705°C. (1300°F.) in flat-rolled bands, finished at 790°C. (1450°F.) with the same cold reductions. No large grains were observed with any reduction or with any annealing temperature in bands finished at 730°C. (1345°F.). Properties for the strip finished at 880°C. (1615°F.) and annealed at 650°C. (1200°F.) were as follows:

Reduction by cold rolling, per cent	Tensile strength, lb. per sq. in.	Yield strength, lb. per sq. in.	Yield ratio, per cent	Elongation, in 2 in., per cent
2	50,300	33,500	66.4	34.0
5	51,000	35,300	69.1	31.0
10	48,500	32,400	66.9	19.0
15	45,000	25,500	56.6	31.0
17.5	46,000	26,800	58.2	30.0

Pomp[753] noticed a coarsening of the grain in a 1.2 per cent carbon steel annealed at 700°C. (1290°F.) after cold drawing with 5 to 15 per cent reduction in area. This coarsening was accompanied by a lowering of the elastic limit from 53,500 to 35,700 lb. per sq. in., and a lowering of the yield strength from 63,900 to 45,200 lb. per sq. in. The lowest values corresponded to reductions in area of about 13 per cent.

According to Brophy and Wyman,[595] the annealing characteristics of cold-worked normal and abnormal steels (normality was judged by the McQuaid-Ehn carburizing test) differ. Figures 44 and 45 show hardness versus annealing-temperature curves for a normal and an abnormal steel respectively. The

* See footnote, p. 102.

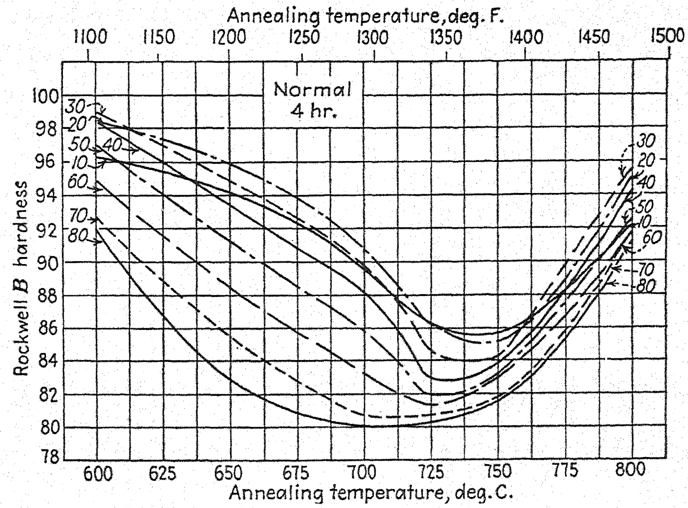

FIG. 44.—Hardness versus annealing temperature for a normal steel containing 0.71 per cent carbon and 0.46 per cent manganese. The numerals on the curves indicate the percentage reduction in cross-section by cold rolling. (*Brophy and Wyman.*[595])

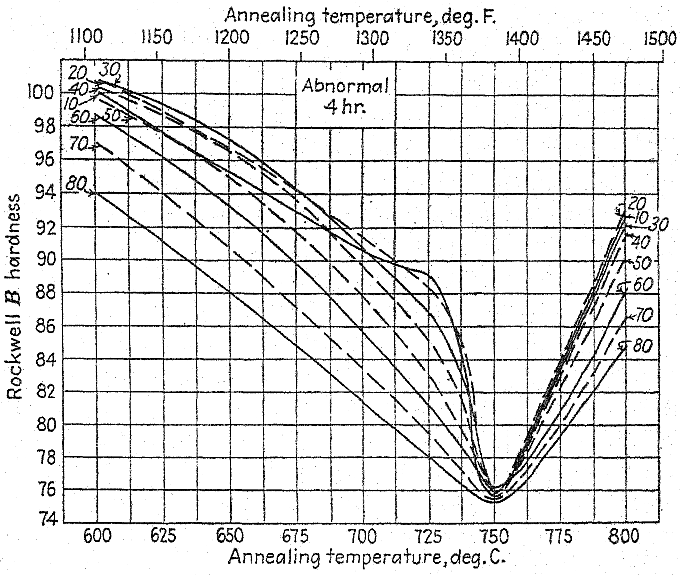

FIG. 45.—Hardness versus annealing temperature for abnormal steel containing 0.73 per cent carbon and 0.45 per cent manganese. (*Brophy and Wyman.*[595])

normal steel contained 0.71 per cent carbon, 0.46 per cent manganese, and 0.07 per cent silicon; the abnormal steel contained 0.73 per cent carbon, 0.45 per cent manganese, 0.07 per cent silicon, and 0.05 per cent chromium. The steels were in the form of strips, and samples were annealed for 4 hr. at the indicated temperatures. Even though the steels had almost identical compositions, they softened to different degrees when annealed at temperatures between 600 and 750°C. (1110 and 1380°F.).

E. AUTHOR'S SUMMARY

1. Iron and steel can be worked by rolling or drawing at or near atmospheric temperature into sections of a variety of shapes and sizes. Many of these sections can be produced more economically by cold working than by any other process of mechanical treatment. Moreover, certain desired properties, including very high tensile strength, which are difficult or even impossible to attain by hot working and subsequent heat treatment can be readily attained by cold working.

2. The principal effect of cold work is to increase strength and hardness and decrease ductility. The change in properties is dependent chiefly upon the amount of cold working.

3. All grades of carbon-steel wire, regardless of the original structural condition, increase in tensile strength by approximately the same amounts with equal amounts of cold work. Irrespective of the original value, the percentage elongation always drops to a low value after approximately 10 per cent reduction by drafting and is affected slightly by further cold working.

4. The tensile strength of hard-drawn wire depends primarily upon the amount of reduction per draft and upon the number of drafts. According to some investigators, the properties are independent of the speed of drawing, the lubricant, or the shape of the die. There is apparently some question about the accuracy of this statement, at least in the case of high-quality wire.

5. The tensile strength of commercial hard-drawn—*i.e.*, with reductions of 85 to 98 per cent without intermediate annealing—low-carbon wire is about 2.5 times the strength of the hot-rolled rod from which it is drawn; for basic open-hearth steel the tensile strength is approximately 125,000 lb. per sq. in., for Bessemer steel it is about 150,000 lb. per sq. in. Elongation for

both grades is usually 2 per cent (in 10 in.) or less. Annealing partially, or in some cases almost completely, recrystallizes the wire and destroys most, or nearly all, of the effects of cold work, resulting in a tensile strength which is as low or lower, and an elongation which is as high or higher, than that of the original hot-rolled rod. One draft after annealing increases the tensile strength 10 to 50 per cent depending upon the size and the percentage reduction by the draft; it decreases the percentage elongation from the high value characteristic of annealed material to 1 to 5 per cent.

6. For cold-drawn wire, cold-rolled rod and strip, and cold-drawn tubes containing 0.10 to 1.40 per cent carbon data are presented which show that the increase in tensile and yield strengths and the decrease in percentage elongation are proportional to the amount of reduction in cold working and are practically independent of the amount of carbon in the steel. Hardness is increased by cold working but does not maintain such a definite relation to the tensile strength as it does in material which is not cold worked.

7. For the same total reduction by cold working, heavy drafts increase the tensile and yield strengths more than light drafts.

8. Finishing temperatures of hot working affect the tensile properties of cold-drawn, and of cold-drawn and annealed, low-carbon strips. Higher tensile strength in the cold-drawn strip results from a finishing temperature of 880°C. (1615°F.) than from finishing temperatures of 730 and 790°C. (1345 and 1450°F.). The yield ratio for strip finished at all three temperatures becomes 100 per cent for 20 to 25 per cent cold reduction. The elongation values (in 2 in.) are lowered by cold rolling to approximately 2 per cent with reductions of about 35 per cent. Further cold working has little effect. In general, the best properties of annealed strip result from hot rolling flat with finishing temperatures of 790°C. (1450°F.) or above. Coiling the hot-rolled strip is apparently responsible for a lower elongation in the annealed sheet. Conclusions are given which correlate the mill and annealing practices with the deep-drawing properties.

9. Wire drawn from medium- and high-carbon rods which have been patented before drawing has a higher tensile strength than wire drawn from unpatented rod. The fine pearlitic structure which results from patenting is more desirable for material which

is to be drawn into high-strength wire than the lamellar pearlite usually present in the unpatented (hot-rolled) rod. The tensile strength of wire drawn from annealed rod is much lower than that of wire drawn from patented rod.

10. High-strength medium-carbon wire drawn from lead-patented rod has approximately the same tensile strength in the larger sizes as wire similarly drawn from air-patented rod. In finer sizes (0.06 in. or smaller) the tensile strength of the wire drawn from lead-patented rod is higher. The improvement in properties caused by lead patenting is even more striking in high-carbon material.

11. Varying the patenting temperature has little effect on the properties of the cold-drawn wire. The safe temperature range for patenting is extremely wide. The temperature of lead into which the wire is quenched does not affect the properties of the cold-drawn wire appreciably so long as it is kept in the range 450 to 550°C. (840 to 1020°F.).

12. Wire drawn from rod which has been oil quenched from about 980°C. (1795°F.) and tempered at 425 to 480°C. (795 to 895°F.) work hardens faster and becomes brittle much sooner than wire drawn from patented rod. The fine structure produced by oil quenching and tempering is quite unsatisfactory for withstanding cold flow.

13. The tensile properties of 0.70 per cent carbon strip cold rolled to 65 per cent reduction in thickness from hot-rolled material are about the same as those of strip similarly rolled from normalized material. If the material is annealed before cold rolling, tensile strength and elongation are lower.

14. If low-carbon steel is of the quality known, for lack of a better name, as "aging steel," cold-worked sections will age harden at atmospheric temperatures, which results in a 10 to 20 per cent increase in tensile and yield strengths and lower values for elongation, reduction of area, and impact resistance. Under the same conditions, the proportional limit (as defined in Chapter I) may be increased 200 to 300 per cent. Data on the relation between the amount of cold working and the effect of aging on the properties are erratic.

15. The changes which, on aging at normal temperature, take place over long periods of time can be greatly accelerated by

heating even to 60°C. (140°F.), and at 100°C. (210°F.) they occur almost immediately.

16. The reheating which takes place when cold-worked steel is galvanized usually decreases the tensile strength and increases the percentage elongation. Galvanizing has an embrittling effect evidenced by lower torsion and bend-test values.

17. Davenport and Bain have shown recently that the amount of hardening caused by strain aging after cold working is probably dependent upon the amount of oxygen in the steel and upon whether or not the material has received a treatment prior to cold working which pre-ages it and reduces its susceptibility to age hardening after cold working. If these conclusions are accepted, the inconsistent results obtained by different investigators on the effect of such variables as the amount of cold working and the like on age hardenability are more easily explainable: the steels differed in oxygen content or in the treatment prior to cold working.

18. The annealing temperatures and the times of holding at temperature which are necessary to remove the hardness caused by cold work depend upon the composition of the steel and the amount of cold work; the temperature decreases as the amount of cold working increases. The annealing characteristics of normal and abnormal steels differ.

19. After annealing at the recrystallization temperature or above, the properties of cold-worked steel are approximately the same as if the material had been hot rolled and annealed. These properties depend chiefly upon the carbon content and are discussed in the next chapter.

CHAPTER VI

MECHANICAL PROPERTIES OF HEAT-TREATED CARBON STEELS

Effect of Normalizing and Annealing—Effect of Quenching—Quenched and Tempered Low- and Medium-carbon Steels—Quenched and Tempered High-carbon Steels—Author's Summary

Statistics for the annual output of heat-treated steels are not published and probably are not compiled at all. From the annual production of rolled steel it may, however, be estimated roughly that less than 10 per cent receives any thermal treatment after hot or cold working except that which takes place spontaneously when medium- and high-carbon steels of relatively small cross-section cool in air from the final rolling or forging temperature. If from this 10 per cent is deducted the alloy-steel production, most of which is heat treated before being used, and the tonnage of wire, sheet, and strip, which is annealed solely to remove the effects of cold work, it is probable that the amount of carbon steel which receives a heat treatment before being used commercially is considerably less than 1 per cent of the total steel production. The economic advantage of heat treatment is, however, great, even when restricted to carbon steels. It is impossible to estimate the increase in the commercial value of certain carbon steels resulting from heat treatment; it gives to these steels, however, a far greater proportion of the total value of rolled-steel production than the above percentage would indicate. The debt which industrial civilization owes to the phenomena included broadly under the term heat treatment has been so vividly pictured by so many writers that further discussion of it is unnecessary.

Methods used for heat treatment of commercial iron-carbon alloys, the principles underlying the various heat-treating operations, and the accompanying constitutional changes have been adequately covered in the first volume of this monograph, nearly half of which is devoted to these subjects. It is only necessary,

therefore, to consider in this second volume the effect of thermal treatment on the properties of carbon steels. For convenience, this discussion has been divided into two parts; the present chapter deals with some of the data on the properties of heat-treated steels of small cross-section (represented principally by the standard tensile-test specimens or small bars from which the test specimens are machined); in the next chapter are data on the effect of various thermal treatments on steels of varying cross-section, in other words, on the effect of mass (see page 215) on the properties.

A. EFFECT OF NORMALIZING AND ANNEALING

Normalizing was defined in Volume I, page 280, as the operation of heating steel to approximately 55°C. (100°F.) above the critical range followed by cooling to below the range in still air at ordinary temperature. Annealing (full annealing) was defined as heating above the critical range and holding above that range for the desired time, followed by slow cooling through the range. As emphasized in Volume I, the difference between normalizing and annealing is primarily a difference in cooling rate and secondarily a difference in the temperature to which the material is heated.

68. Properties of Normalized Carbon Steels.—A large part of the hot-worked steel produced is actually in the normalized condition when cold; the finishing temperature for rolled or forged sections is often 40 to 60°C. (70 to 110°F.) above the critical range, and the cooling is nearly always in still air. Thus, the properties of a normalized steel should differ but little from the properties of a hot-rolled steel of corresponding composition.

The tensile properties of a series of iron-carbon alloys of fairly high purity were determined by Arnold.[3] The compositions were as shown in the table on page 174.

Arnold melted 50-lb. heats in a crucible, cast them into 3 × 3-in. ingots, and hammered and rolled the ingots to $1\frac{1}{16}$-in. round bars. As the manganese and silicon contents were purposely kept at a minimum, the alloys were deoxidized with aluminum; all contained appreciable amounts of this element, probably in form of the oxide. Sections of the hot-rolled bars, heated to 1000°C. (1830°F.) and cooled in air, were machined to

test pieces 0.564 in. in the reduced section with a 2-in. gage length. The tensile properties are plotted in Fig. 46. The value

Element, per cent					
C	Mn	Si	P	S	Al
0.08	0.02	0.03	0.02	0.03	0.02
0.21	0.05	0.05	0.02	0.03	0.02
0.38	0.08	0.03	0.02	0.02	0.03
0.59	0.10	0.07	0.02	0.02	0.03
0.74	0.01	0.05	0.02	0.02	0.02
0.89	0.09	0.03	0.02	0.02	0.03
1.20	0.15	0.07	0.02	0.02	0.03
1.47	0.13	0.08	0.02	0.01	0.04

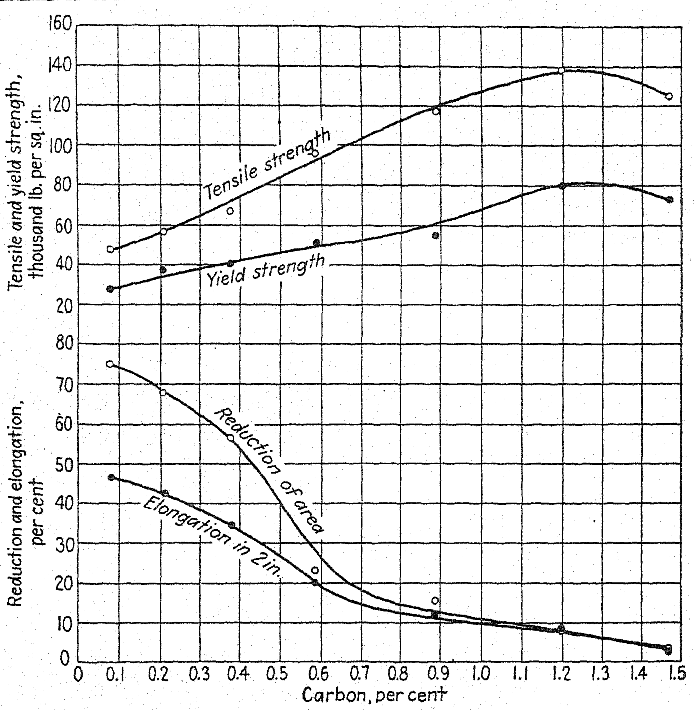

FIG. 46.—Tensile properties of normalized iron-carbon alloys of high purity. (*Arnold.*[3])

designated yield strength in Fig. 46 was reported by Arnold as yield point.

The tensile properties of a series of normalized acid open-hearth steels as determined by Brinell and reported by Wahlberg[5] are plotted in Fig. 47. Specimens, 32 mm. (1.26 in.) in diameter, were heated to 1000, 850, or 750°C. (1830, 1560, or 1380°F.) and cooled in air. Test bars were 18 mm. (0.71 in.) in

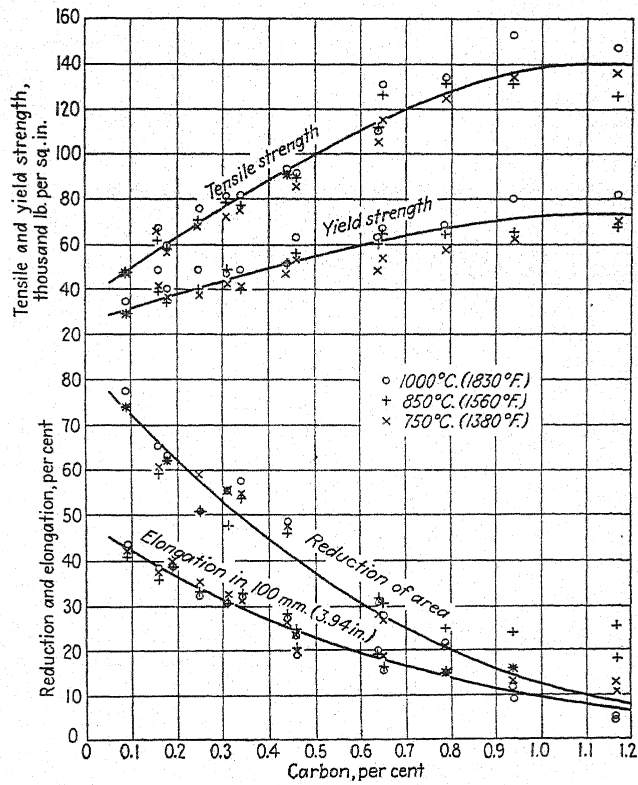

Fig. 47.—Tensile properties of acid open-hearth steels, normalized from the indicated temperatures. (*Brinell according to Wahlberg.*[5])

diameter with a 100-mm. (3.94-in.) gage length. As shown by Fig. 47, slightly higher tensile- and yield-strength values resulted from the higher normalizing temperatures. Brinell used only three temperatures; consequently he did not detect the dip in strength and the peak in elongation or reduction of area which occur in high-carbon steels when normalized or annealed at 700 to 900°C. (1290 to 1650°F.), and which are shown by the

results of Meyer and Wesseling[166] discussed in section 70, page 179. Incidentally, these last-named results do not wholly confirm the conclusion of Brinell, at least for high-carbon steels, that slightly higher tensile and yield strengths result from the higher normalizing temperatures. As shown in Figs. 50 and 51 (pages 181 and 182), tensile and yield strengths were higher after normalizing at 850°C. (1560°F.) than they were after normalizing at 750°C. (1380°F.), but there was little if any difference in the strength properties after normalizing at 850 and 1000°C. (1560 and 1830°F.).

If the curves of Fig. 47 and of Fig. 10 from Chapter IV, page 85, for hot-rolled steels are superimposed, it will be noted that the tensile- and yield-strength values for normalized commercial steels check fairly well with those for the corresponding hot-rolled steels. The values of elongation and reduction of area are higher than the values for hot-rolled steels of the same carbon content. Arnold's tensile- and yield-strength values (Fig. 46) are markedly lower than the values of Brinell. The slight differences in elongation and reduction of area between the values of Arnold, Brinell, and those given in Fig. 10 are not indicative of any definite effect of normalizing or of purity because of the difference in gage length and the scatter of Brinell's points and of those shown in Fig. 10.

The data plotted in Fig. 49, page 180, also show how closely the mechanical properties of the rolled and normalized materials may check. The actual values for the normalized material, plotted in Fig. 49, do not fall close to the curves for normalized alloys in Fig. 47, especially in the low- and medium-carbon ranges. This is not surprising as the values in Fig. 49 are typical averages for ranges of carbon, while in Fig. 47 the plotting points represent actual values for alloys of definite composition.

It might be tentatively concluded, subject to modification when more data become available, that: (*a*) normalizing commercial hot-worked carbon steels improves the elongation and reduction of area and has little or no effect on the tensile and yield strengths; and (*b*) normalized carbon steels of fairly high purity (deoxidized with aluminum) have markedly lower tensile and yield strengths (and, perhaps, in the medium-carbon ranges higher elongation and reduction of area) than normalized commercial carbon steels.

69. Effect of Carbon on Properties of Annealed Steels.—The effect on mechanical properties of annealing (process annealing) at temperatures below the critical, used chiefly for softening cold-worked steels, was discussed briefly in the previous chapter; in the present section and the following one, attention has been paid

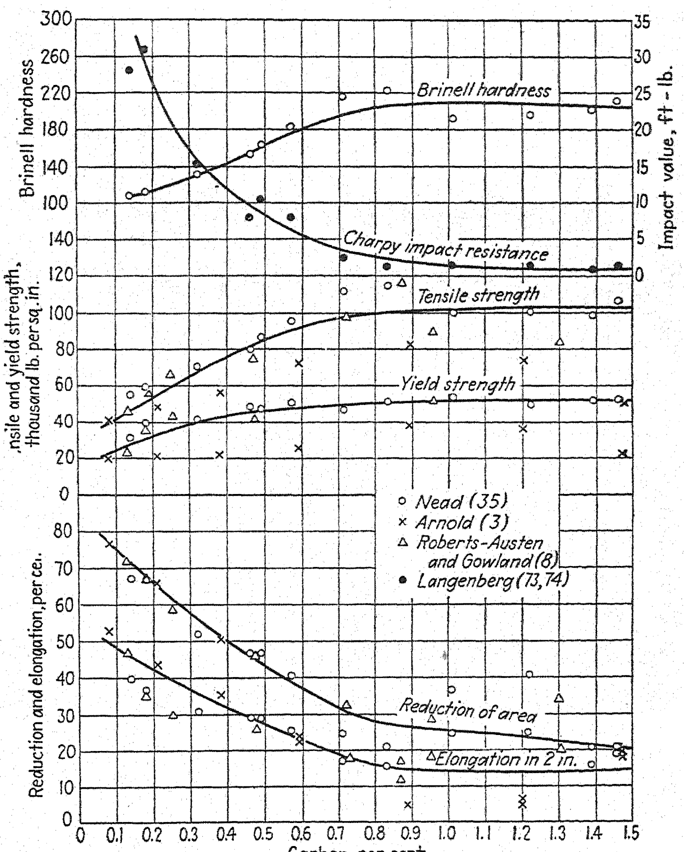

FIG. 48.—Effect of carbon on properties of annealed steels.

to the effect on properties of annealing at temperatures above the critical range. As noted on page 173, this is known as "full annealing" or, in the mill, as "annealing."

The tensile properties of annealed steels of varying carbon content plotted from the data of four investigators are given in Fig. 48. The complete compositions of the alloys used by

Nead[35] and Langenberg[73,74] are tabulated on page 84, Chapter IV; carbon contents and annealing temperatures were as follows:

Carbon, per cent	Annealing temperature*	
	°C.	°F.
0.14	866	1590
0.18	858	1575
0.32	836	1525
0.46	819	1505
0.49	816	1500
0.57	809	1490
0.71	800	1470
0.83	795	1465
1.01	790	1455
1.22	790	1455
1.39	790	1455
1.46	790	1455

* Also used as quenching temperatures in investigations discussed later in this chapter.

The specimens were heated to the proper temperature and were cooled slowly in a small muffle.

The steels used by Arnold were the fairly pure alloys whose compositions are given on page 174; samples were heated to 1000°C. (1830°F.) for 72 hr. and "cooled for 100 hr. in a luted furnace." The steels whose properties were given in the *Report* to the Alloys Research Committee[8] were commercial alloys; samples were heated to 800°C. (1470°F.) and allowed to cool in the muffle. All of the data on tensile properties shown in Fig. 48 were obtained with samples approximately 0.5 in. in diameter having a 2-in. gage length. Arnold's values of yield strength and some of his values of tensile strength are surprisingly low. This may be attributed to the purity of his alloys, to extremely slow cooling from the annealing temperature, to the deoxidation by aluminum, or to his method of testing.

In Fig. 49 are charted the properties of four carbon steels, determined by the Bethlehem Steel Company,[791] as rolled, normalized (air cooled), and annealed (furnace cooled). The steels and the temperatures of treatment are as follows:

S.A.E. number	Carbon range, per cent	Manganese range, per cent	Normalizing temperature		Annealing temperature	
			°C.	°F.	°C.	°F.
1015	0.10 to 0.20	0.30 to 0.60	925	1700	870	1600
1040	0.35 to 0.45	0.60 to 0.90	900	1650	790	1450
1080	0.75 to 0.90	0.30 to 0.50	900	1650	790	1450
1095	0.90 to 1.05	0.25 to 0.50	900	1650	790	1450

The test pieces were standard 0.505-in. diameter specimens with a 2-in. gage length and were machined after treatment from a 1-in. round bar. Prefacing these results, the Bethlehem Steel Company stated: "These data represent average physical properties. They are not the maximum obtainable, nor the minimum which may be anticipated."

The values in Fig. 49 do not check with the values plotted in Fig. 48. The discrepancy is especially apparent in the values for the tensile strength of the 1015 and 1040 steels, which are about 10,000 lb. per sq. in. higher in Fig. 49, and in the yield strength, which in Fig. 49 is 5,000 to 10,000 lb. per sq. in. higher throughout the whole range of carbon than in the composite curves in Fig. 48. Despite the fact that the Bethlehem Steel Company's strength values are markedly higher than the average values given in Fig. 48, the values for reduction of area are also considerably higher, as much as 75 per cent in the upper carbon ranges. It is, of course, true that the data plotted in Fig. 48 are actual determinations made on steels of definite carbon content, while the values plotted in Fig. 49 are average properties of steels with a range of carbon. Even so, the discrepancy is too great and should be narrowed by further work.

70. Effect of Annealing Practice on Properties of High-carbon Steels.—The effect of annealing practice on the properties of carbon steels containing 0.74 to 1.15 per cent carbon was studied by Meyer and Wesseling.[166] Specimens (size and gage length not given) were treated as follows:

A. Annealed 30 min. and air cooled.
B. Annealed 5 hr. and air cooled.
C. Annealed 30 min. and furnace cooled.
D. Annealed 5 hr. and furnace cooled.

Impact tests were made on samples with a keyhole notch and a cross-section at the point of fracture of 15 by 10 mm. (0.59 by 0.39 in.).

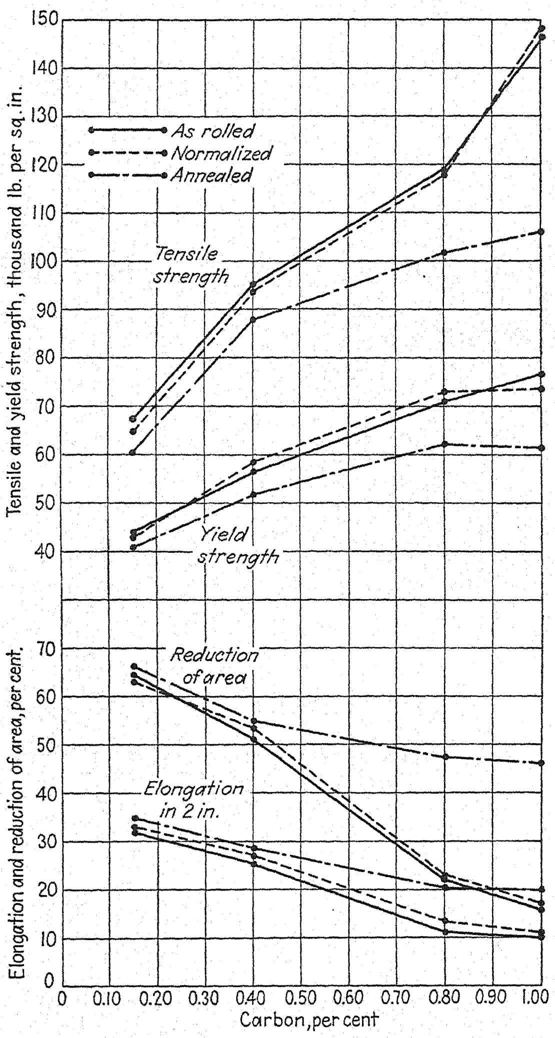

FIG. 49.—Properties of rolled, normalized, and annealed carbon steels. (*Bethlehem Steel Company.*[791])

Results are plotted in Figs. 50 and 51; the former gives values for the 0.74 and 0.935 per cent carbon steels, the latter the values

for the higher carbon specimens. Strictly speaking, treatments *A* and *B* should be termed normalizing; hence, the curves in these two figures show the effect of the normalizing or annealing temperature, the effect of air versus furnace cooling, and,

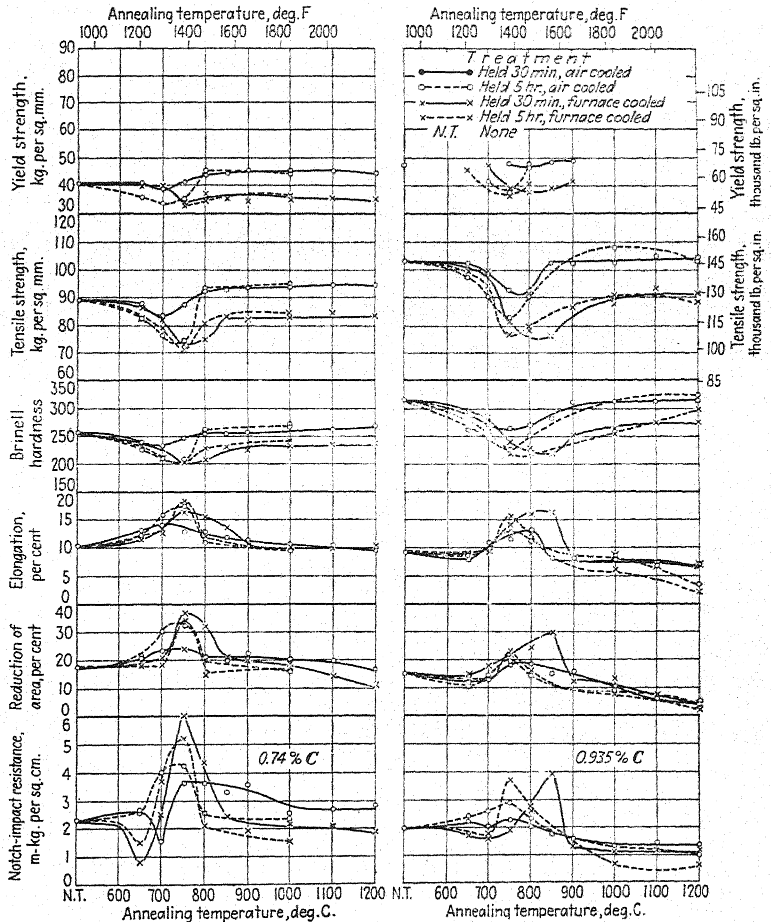

FIG. 50.—Effect of annealing practice on the mechanical properties of 0.74 and 0.935 per cent carbon steels. (*Meyer and Wesseling.*[166])

to some extent, the effect of the time of holding at the required temperature.

The outstanding feature of Meyer and Wesseling's results is the dip in tensile strength, yield strength, and hardness and the

rise in elongation, reduction of area, and impact resistance when the annealing and normalizing temperatures exceeded the critical —about 750°C. (1380°F.)—and the complete or partial return

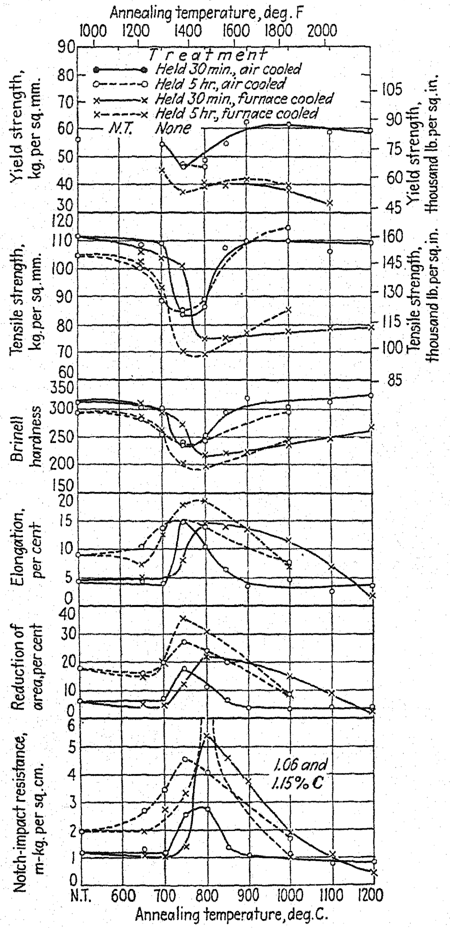

FIG. 51.—Effect of annealing practice on the mechanical properties of 1.06 and 1.15 per cent carbon steels. (*Meyer and Wesseling.*[166])

of these properties to their original value when the specimens were annealed or normalized at 900°C. (1650°F.) or above.

It is interesting to note that, considering the uncertainty of the exact composition of the Bethlehem Steel Company steels (Fig. 49) and the differences in specimen size, the properties

obtained by Meyer and Wesseling for a normalizing temperature of 900°C. (1650°F.) and an annealing temperature of 800°C. (1470°F.) fall fairly close to the Bethlehem Steel Company curves for steels of approximately the same composition.

There are two other tentative conclusions which may be drawn from the results in Figs. 50 and 51. These are: (1) In the case of the 0.74 and 0.935 per cent carbon steels, there seems to be little if any change in the properties when the time of holding at temperature is increased from 30 min. to 5 hr.; and (2) the conclusion drawn on page 176 that the tensile strength of hot-worked steel is affected little if any by normalizing is valid only if the normalizing temperature is 900°C. (1650°F.) or above (as is usually the case in practice) and possibly only if the carbon is less than 1 per cent.

B. EFFECT OF QUENCHING

The principles underlying the hardening of iron-carbon alloys are discussed at length in Chapter VI of Volume I of this monograph, and quenching media and internal stresses in Chapter X. Because of the severe stresses introduced when the heat is extracted rapidly, and incident to the formation of martensite, steels containing enough carbon to make heat treatment advantageous are rarely, if ever, used in the quenched and untempered condition. It is, however, of interest academically, if not practically, to correlate the few data available on the effect of quenching on the properties of commercial iron-carbon alloys.

As mentioned at the beginning of this chapter, the properties given hereafter were obtained on specimens of small cross-section, about 1 in. in diameter or usually less; hence, these properties of the quenched steels are dependent chiefly upon the cooling rate; mass, because of its virtual constancy, is a negligible or unimportant factor.

71. Quenched Low-carbon Steels (Less than 0.15 Per Cent Carbon).—The question whether pure iron can be hardened by quenching is not yet answered definitely. This has been discussed in a previous monograph[794] (page 399) and in Chapter VI of Volume I of this monograph. Commercially pure iron, which in reality is a fairly pure iron-carbon alloy containing 0.01 to 0.03 per cent carbon, can unquestionably be hardened by quench-

ing. This is shown by Kenyon, from whose paper[261] the properties in Table 41 were taken.

Hanemann and Kühnel[26] reported on the effect of quenching and tempering a steel containing 0.05 per cent carbon, 0.46 per cent manganese, 0.006 per cent silicon, 0.04 per cent sulphur, and 0.06 per cent phosphorus. To facilitate a rapid heat extraction, specimens 8.9 mm. (0.35 in.) in diameter were used. Elongation

TABLE 41.—EFFECT OF QUENCHING ON THE PROPERTIES OF BASIC OPEN-HEARTH INGOT IRON*

Property	Annealed 25 hr. at 900°C. (1650°F.)	Quenched in water from 940°C. (1725°F.)
Tensile strength, lb. per sq. in.	41,000	47,000
Yield strength, lb. per sq. in.	18,300	30,300
Elongation in 2 in., per cent	47.0	36.2
Reduction of area, per cent	70.6	70.0
Brinell hardness	82	110
Impact value, ft-lb. per sq. in.	695	452

* Kenyon.[261]

was measured over a gage length of 100 mm. (3.94 in.); hence, the reported values will be lower than elongation values on a United States standard specimen with a 2-in. gage length. Properties after the various quenching treatments are given in Table 42.

TABLE 42.—EFFECT OF QUENCHING ON THE PROPERTIES OF A STEEL CONTAINING 0.05 PER CENT CARBON*

Treatment	Tensile strength, lb. per sq. in.	Elongation in 100 mm. (3.94 in.), per cent	Reduction of area, per cent
Annealed at 800 to 900°C. (1470 to 1650°F.)	52,200	26.0	75.0
Quenched in oil from 750°C. (1380°F.)	80,000	16.6	61.7
Quenched in oil from 850°C. (1560°F.)	71,200	14.0	58.0
Quenched in oil from 950°C. (1740°F.)	76,300	11.4	61.0
Quenched in water from 850°C. (1560°F.)	77,300	13.5	55.1
Quenched in water from 950°C. (1740°F.)	102,500	11.0	51.0

* Hanemann and Kühnel.[26]

An investigation conducted by a Committee of the American Society for Testing Materials[79] to determine the effects of sulphur on the properties of rivet steel is of interest here as it indicated the response of low-carbon steels to heat treatment. Fourteen basic open-hearth steels were tested, but for the present purpose it will be sufficient to show the properties of only one of them, which may be considered as representative of all fourteen. The steel contained 0.11 per cent carbon, 0.45 per cent manganese, 0.01 per cent silicon, 0.044 per cent sulphur, and 0.008 per cent phosphorus. It was hot rolled to $^{63}/_{64}$-in. rounds and heat treated before machining. The properties of the steel in the hot-rolled condition, as annealed between 925 and 950°C. (1695 and 1740°F.), and as quenched in water from the same temperature range are listed in Table 43. The quenching treatment increased the tensile strength from 49,000 to 63,000 lb. per sq. in. This increase is considerably less than that found by Hanemann and Kühnel for a steel containing only 0.05 per cent carbon, which may be attributable to the use of a larger specimen, a difference in the steels not obvious from the reported analyses, or a less drastic quench.

TABLE 43.—EFFECT OF ANNEALING AND QUENCHING ON THE PROPERTIES OF A STEEL CONTAINING 0.11 PER CENT CARBON*

Property	Rolled	Annealed	Quenched
Tensile strength, lb. per sq. in.	49,500	49,000	63,000
Proportional limit, lb. per sq. in.	34,000	26,000	39,500
Yield strength,† lb. per sq. in.	35,500	26,000	43,000
Elongation in 2 in., per cent	46	47	34
Reduction of area, per cent	73	73	74
Charpy impact value, ft-lb	57	9	60
Izod impact value, ft-lb	119	34	112.5
Brinell hardness	94	86	136
Scleroscope hardness	16.5	16	19

* A.S.T.M. Committee.[79]
† Reported as yield point.

In an investigation of the effect of quenching on low-carbon steels, Smith[139,170] found that a steel containing approximately 0.10 per cent carbon can be appreciably hardened by a drastic quench and that such a treatment is commercially feasible for the production of high-strength bolts. He used a special appara-

tus in which the material could be rapidly cooled by means of a water spray. The solid lines in Fig. 52 show the tensile strength and reduction of area of standard A.S.T.M. specimens of a 0.13 per cent carbon steel drastically quenched from various tempera-

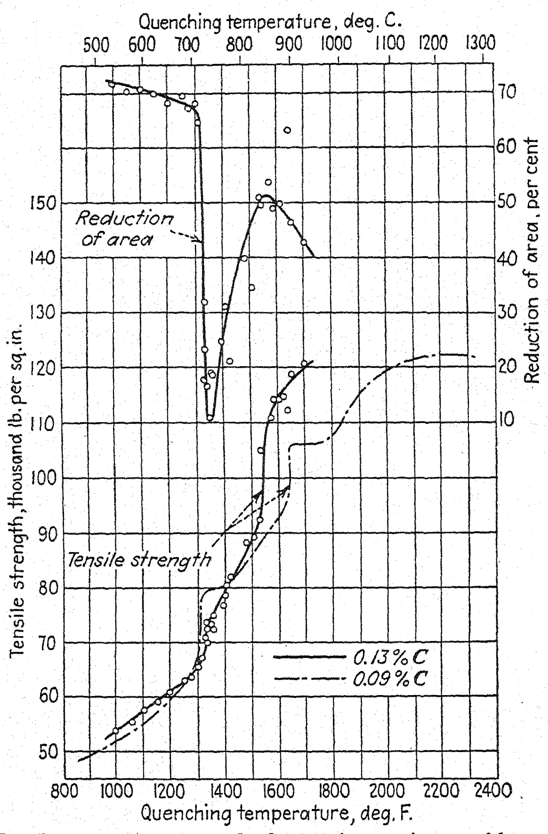

FIG. 52.—Tensile properties of standard 0.505-in. specimens of low-carbon steel drastically quenched in a water spray. (*Smith.*[139])

tures; the dotted line shows the tensile strength of a steel containing 0.09 per cent carbon. For a certain range immediately above the lower critical temperature the strength increased almost linearly with the quenching temperature. Low values of reduction of area resulted from quenching near the lower critical temperature. Smith's data also indicate that the steels can be strengthened by quenching from below the lower critical tempera-

ture. If this increase in hardness is real, it can probably be attributed to precipitation-hardening phenomena.

72. Quenched Steels of Varying Carbon Content.—Nead(35) and Langenberg,(73,74) whose data on the properties of hot-rolled and annealed carbon steels have been quoted in previous sections, also determined the properties of the same series of carbon steels after oil and water quenching. The quenching temperatures were the same as the annealing temperatures given on page 178. Specimens were approximately 0.5 in. in diameter with a 2-in. gage length and were untempered.

In connection with its work on the mechanical properties of automobile steels, the British Steel Research Committee(55) reported properties after water and oil quenching and before tempering. Among the carbon steels used were the following:

Composition, per cent					Quenching temperature	
C	Mn	Si	S	P	°C.	°F.
0.12	0.71	0.05	0.080	0.048	900	1650
0.17	0.72	0.24	0.041	0.047	920	1690
0.26	0.75	0.30	0.044	0.050	900	1650
0.45	0.78	0.32	0.020	0.025	870	1600

The specimens used by the Steel Research Committee were British standard specimens 0.564 in. in diameter with a 2-in. gage length. The results which are reported in Fig. 53 and Table 44 as yield strength were determined as yield point. As all of the experimental conditions of heat treatment, preparation of specimens, and method of testing were carefully standardized and as the results averaged by the Committee were obtained under this standard procedure by three laboratories working independently of each other, there is reason to believe that the values as reported in this and following sections represent accurately the properties of steels of the composition given for the heat treatments used.

A summary of the data reported by Nead, Langenberg, and by the Steel Research Committee is given in Fig. 53 for the untempered oil-quenched steels and in Table 44 for the untempered water-quenched steels. As shown in the former, tensile

strength, yield strength, and Brinell hardness increase linearly with the carbon increasing from 0.1 to approximately 0.7 per cent. The Brinell hardness values given in Fig. 53 indicate that oil quenching is not drastic enough for complete hardening.

According to the Brinell values obtained by Nead, water quenching is drastic enough for complete hardening; but the

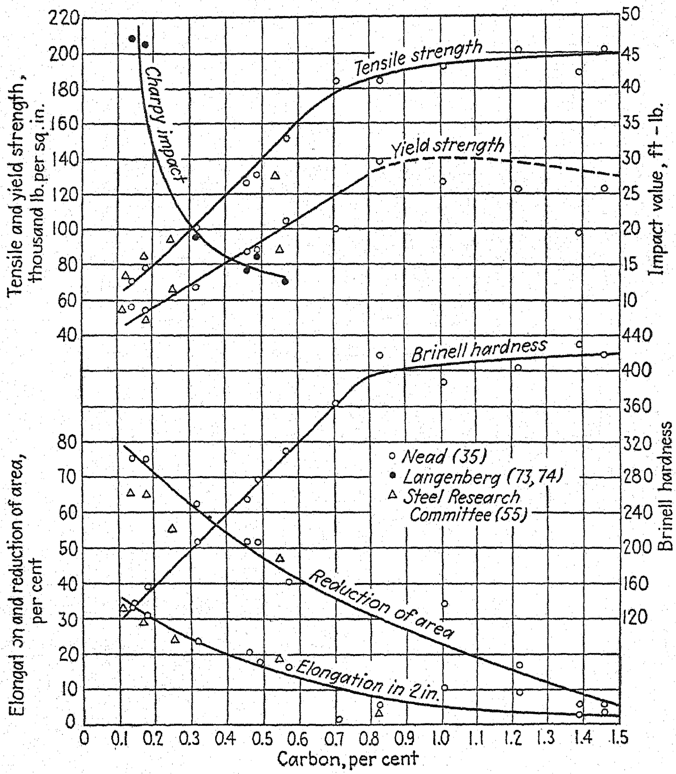

FIG. 53.—Effect of carbon on the tensile properties of oil-quenched steels.

internal stresses and probably internal cracks resulting from water quenching the test specimens containing more than 0.40 per cent carbon rendered the tensile values of these higher carbon steels of such doubtful value that they were not reproduced as a part of Table 44. The specimens tested by the Steel Research Committee were machined from quenched bars 1⅛ in. in diameter, which accounts for the lower tensile strength and hardness and the higher elongation and reduction of area as compared

with Nead's results from water-quenched specimens which were but slightly larger than the test piece. The larger mass of the Steel Research Committee's water-quenched specimens had a marked effect on the properties.

TABLE 44.—PROPERTIES OF CARBON STEELS QUENCHED IN WATER

Investigator	Carbon, per cent	Tensile strength, lb. per sq. in.	Yield strength,* lb. per sq. in.	Elongation in 2 in., per cent	Reduction of area, per cent	Brinell hardness	Izod impact, ft-lb.
Nead[35] and Langenberg[73,74]	0.14	90,000	....	21.0	67.0	170	38.1†
	0.18	105,000	....	16.5	57.2	228	35.7†
	0.32	135,000	....	8.0	16.9	255	10.9†
	0.46	220,000	....	1.0	Nil	600	
Steel Research Committee[55]	0.12	89,600	67,200	26	59	183	48
	0.17	103,000	78,400	22	51	223	24
	0.26	127,700	80,600	11	30	...	13
	0.45	150,000	107,500	12	28	321	14

* Reported as yield point. As the yield-strength values obtained by the different laboratories varied widely, these results are only approximate.

† Charpy impact, ft-lb.

The presence of internal microscopic cracks which affect adversely the mechanical properties of drastically quenched high-carbon steels has been recently demonstrated by Davenport, Roff, and Bain.[684] The microcracks or their causes cannot be removed by ordinary tempering; these investigators found that it was necessary in one instance to reheat the specimen to 1090°C. (2000°F.) to effect their complete elimination. As is shown on page 206, a 0.74 per cent carbon steel, treated so that this form of cracking is avoided, has higher tensile strength, yield strength, elongation, reduction of area, and impact resistance than the same steel oil quenched and tempered in the usual manner to the same hardness.

Because of the known presence in drastically quenched steel of very high internal stresses, and because of the probable presence of microcracks, the values for tensile strength, yield strength, elongation, and reduction of area for the untempered oil-quenched steels containing more than 0.70 per cent carbon, shown in Fig. 53 (as determined on standard test pieces heat

treated in the form of ¾-in. or smaller bars), are probably too low. For the untempered water-quenched steels, the results of which are given in Table 44, these values (especially as obtained by Nead on similarly small quenched bars) are undoubtedly too low, except possibly for the steels containing less than 0.20 per cent carbon.

C. QUENCHED AND TEMPERED LOW- AND MEDIUM-CARBON STEELS

Low-carbon steels (less than 0.15 per cent carbon) are seldom quenched and tempered; except for a few special applications—for example high-strength bolts (see page 185)—practically all of the low-carbon steels are used in the hot-rolled or, in the case of strip, wire, and allied products, in the annealed condition. A few of the structural and forging steels (0.16 to 0.60 per cent carbon) are quenched and tempered, chiefly to improve the ductility. Within certain limits, the strength of hot-rolled carbon steel increases with the carbon content; there is a corresponding but greater decrease in elongation and reduction of area. Thus, as the carbon increases from 0.10 to 0.80 per cent, the tensile strength of hot-rolled steel increases to 2 to 2.5 times its original value, but the elongation decreases to about one-third and the reduction of area to about one-sixth of the original values. Structural and forging steels, therefore, may be quenched and tempered, not because the desired strength cannot be attained in untreated material, but because untreated steels of comparatively high strength are not sufficiently tough and can be made tougher by the proper heat treatment.

73. Mechanical-property Charts.—Charts showing the effect of tempering at increasing temperatures on the mechanical properties of carbon and alloy steels have been prepared by the British Steel Research Committee,[55] the U. S. Army Air Corps,* the American Society for Metals,[821] French and Sands of the International Nickel Company,[700] the Society of Automotive Engineers,[817] the Bethlehem Steel Company,[791] and others.

In some instances the properties indicated by the chart are probably the average of a large number of values, in others the average of a few determinations, and in others the minimum

* Printed by U. S. Army Air Corps, Wright Field, Dayton, O., for restricted distribution.

values which may be expected from test bars of standard size heat treated under standard conditions.

Regarding the use of such charts, the "S.A.E. Handbook"(S17) states:

> These . . . physical-property charts should not be considered in any way . . . as giving absolute values. They are intended only as a guide to proper heat treatment of the steels and to indicate conservative physical properties that may be expected of standard test specimens 0.505 × 2 in., machined from rolled bars up to 1.5-in. diameter or square.

The properties given in these charts are ordinarily subject to a number of errors. First and most important: the values are not for a steel of specific analysis but for steels of a composition range. Thus, the chart for S.A.E. 1045 will attempt to give representative properties after heat treatment of steels containing 0.40 to 0.50 per cent carbon and 0.50 to 0.80 per cent manganese. It is hardly necessary to state that after water quenching and tempering at any definite temperature, the properties of a steel containing 0.40 per cent carbon and 0.50 per cent manganese will vary to a marked degree from the properties of a steel containing 0.50 per cent carbon and 0.80 per cent manganese. Although the charts, if properly prepared, will give properties which should be close to the mean of the properties of the above extremes, the chances for error are relatively great. Second: errors may be introduced because of inherent differences in steels of the same composition, made by different processes, or in different plants, with different degrees of deoxidation and freedom from visible and invisible impurities, and with different tendencies for grain growth. Third: the results may be affected by marked differences in heat-treatment practice, such as variations in cross-section of the quenched piece (even if the charts apply only to sections smaller than 1.5 in.), time of holding at temperature, furnace gradients, character, temperature, and circulation of the quenching medium, errors in pyrometers, and many others. Fourth: errors may be introduced by variations in the technique of testing.

The second, third, and fourth sources of error enumerated in the preceding paragraph are, of course, only of marked importance when comparing property charts for steels of different

origin. If the charts are plotted from mechanical properties determined by a single laboratory on steels made by the same practice in the same plant and heat treated under the same conditions, the composition variable is the only one which has an important effect on results. Such charts, while sufficiently trustworthy to fix the average properties of the steels made in that plant, may not, however, indicate with sufficient accuracy the mechanical properties, after specified treatments, of all steels of that particular composition range.

Notwithstanding the many sources of errors in the data given in the published charts, these are undoubtedly of value in indicating the mechanical properties of steels of median composition which will, *in the majority of cases,* result from each specific heat treatment. Hence, in the following pages new property charts have been prepared for carbon structural and forging steels where there were sufficient data available. For each chart the data have been weighed, if necessary averaged, and the properties indicated by a hatched area instead of the customary line. This has the advantage of indicating not only the properties which may be expected from a specific heat treatment, but also the possible variation which may result from each tempering treatment.

The application of statistical methods to the study of properties of heat-treated carbon and alloy steels is urgently needed. The Society of Automotive Engineers [817] has prepared probability curves of the properties of a heat-treated nickel-chromium (3130) and a chromium-vanadium (6130) structural steel, but, as far as could be discovered in the literature, this has not been done for carbon steels or for other alloy steels. In the case of 6130 material, at each tempering temperature the probability curves indicate that there is a fairly constant variation of 15,000 to 18,000 lb. per sq. in. between the maximum or minimum tensile strength and the corresponding value of greatest frequency for any tempering temperature. In the curves for yield point, reduction of area, and Rockwell and scleroscope hardness the spread between the maximum or minimum and the curve of greatest frequency is greater for low tempering temperatures. In the curves given in Figs. 54 to 56 (pages 195, 199, and 200), which were not prepared from a large number of individual values analyzed statistically but from the correlation of a few property curves, each of which may represent average or minimum values,

the same sort of spread is indicated in several instances. If the original property charts from which Figs. 54 to 56 were prepared are of reasonable accuracy, the mechanical properties of 1020, 1035, and 1045 steels, if of average composition and heat treated under fairly constant conditions, should fall within the hatched areas 95 per cent of the time.

74. Quenched and Tempered Low-carbon Steels (Less than 0.15 Per Cent Carbon).—Kenyon[261] found that tempering reduced the strength, hardness, and impact resistance and increased the elongation of water-quenched ingot iron containing 0.01 to 0.03 per cent carbon. Typical values are given in Table

TABLE 45.—EFFECT OF TEMPERING ON MECHANICAL PROPERTIES OF BASIC OPEN-HEARTH INGOT IRON*

Property	Quenched in water from 940°C. (1725°F.)	
	Un-tempered	Tempered at 650°C. (1200°F.)
Tensile strength, lb. per sq. in.	47,000	42,700
Yield strength, lb. per sq. in.	30,300	20,800
Elongation in 2 in., per cent	36.2	41.8
Reduction of area, per cent	70.0	71.1
Brinell hardness	110	94
Impact value, ft-lb. per sq. in.	452	276

* Kenyon.[261]

45. Yield strength and impact resistance were lowered to a marked degree.

The effect of tempering on a 0.05 per cent carbon steel, drastically water quenched from 950°C. (1740°F.), (see page 184 for composition) was studied by Hanemann and Kühnel.[26] As shown by Table 46, the tensile strength dropped rapidly from 103,000 lb. per sq. in. for the quenched specimen to 82,000 lb. per sq. in. for the specimen tempered at 100°C. (210°F.) and to less than 70,000 lb. per sq. in. for the specimens tempered at 200°C. (390°F.) or above. As these tests were made on specimens 8.9 mm. (0.35 in.) in diameter, the results are of course valid only for such small specimens treated in the same way.

Typical properties after tempering at various temperatures of 0.12 and 0.13 per cent carbon steels with 0.41 and 0.71 per cent

TABLE 46.—EFFECT OF TEMPERING ON THE MECHANICAL PROPERTIES OF DRASTICALLY WATER-QUENCHED 0.05 PER CENT CARBON STEEL*

Tempering temperature		Tensile strength, lb. per sq. in.	Elongation in 100 mm. (3.94 in.), per cent	Reduction of area, per cent
°C.	°F.			
None		102,500	11.5	51
100	210	82,000	11.5	63
200	390	64,000	11.5	76
300	570	68,500	11.0	75
400	750	67,500	12.0	75

* Hanemann and Kühnel.(26)

manganese were determined by the Steel Research Committee.(55) The complete analysis of the two steels was as follows:

Element	Percentage	
	A	*B*
Carbon	0.13	0.12
Manganese	0.41	0.71
Silicon	0.06	0.05
Sulphur	0.025	0.080
Phosphorus	0.036	0.048
Nickel	0.20	Nil
Chromium	0.08	Nil

Bars 1¼ in. in diameter, rolled from the bottom half of the ingots were rough machined to 1⅛ in. and sent to three laboratories for testing.

As noted on page 187, all testing procedures, including heat treatment, size and preparation of specimen, and method of testing, were carefully standardized. The average results from each laboratory were correlated and analyzed by the Committee. The values finally reported were those which, "deduced from the investigator's results, were considered by the Committee to represent the mechanical properties of the steel after the heat treatment specified." The results are given in Table 47; as the

work was done with great care, these values may be considered to represent accurately the effect of tempering on a 0.12 per cent carbon steel with low and medium manganese contents.

75. Quenched and Tempered Structural Steels (0.16 to 0.30 Per Cent Carbon).—Published data on the mechanical properties of quenched and tempered 0.16 to 0.30 per cent carbon steels are few. This is to be expected as most of the carbon steels in this

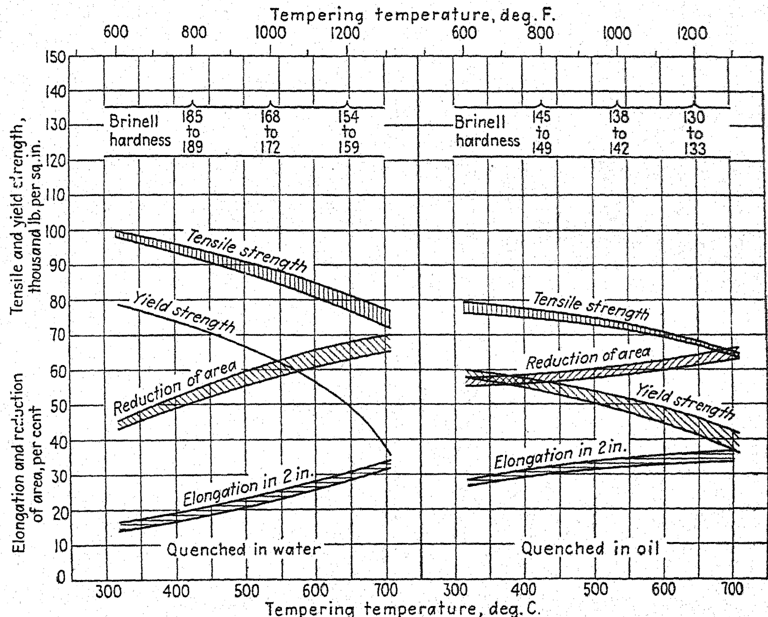

FIG. 54.—Effect of tempering on the tensile properties of 0.15 to 0.25 per cent carbon, 0.30 to 0.60 per cent manganese (1020) steel. Sections, 0.5 to 1.50 in. in diameter, were quenched in oil and water from 870 to 900°C. (1600 to 1650°F.) and tempered (about 60 min.) as shown.

range, especially those containing less than 0.25 per cent carbon, are used primarily for carburizing, and others, containing 0.20 to 0.30 per cent carbon, are the common structural materials nearly all of which are used in the hot- or cold-rolled condition.

From the limited data, including property charts by French and Sands[(700)] on water- and oil-quenched S.A.E. 1020 steel, by Nead[(35)] on oil-quenched material containing 0.18 per cent carbon (see page 84 for complete analysis), and by the U. S. Army Air Corps* on water-quenched S.A.E. 1025 steel, the charts shown

* See footnote p. 190.

in Fig. 54 were constructed. The only values available for the yield strength of the water-quenched material were those of French and Sands shown by a single line in Fig. 54. These charts indicate that the best combination of properties for heat-treated material containing approximately 0.20 per cent carbon and 0.30 to 0.60 per cent manganese is secured by quenching in water and tempering at 550 to 650°C. (1020 to 1200°F.). Brinell-hardness values obtained by these investigators lie on a curve which parallels closely the tensile strength.

TABLE 47.—EFFECT OF TEMPERING ON MECHANICAL PROPERTIES OF A WATER-QUENCHED 0.12 PER CENT CARBON STEEL CONTAINING *A*—0.41 PER CENT MANGANESE AND 0.036 PER CENT PHOSPHORUS AND *B*—0.71 PER CENT MANGANESE AND 0.048 PER CENT PHOSPHORUS*

Steel	Tempering temperature		Tensile strength, lb. per sq. in.	Yield strength,† lb. per sq. in.	Elongation in 2 in., per cent	Reduction of area, per cent	Brinell hardness	Impact value, ft-lb.‡	
	°C.	°F.						Izod	Charpy
A	None		76,200	§	33	71	149	103	21.5
	300	570	76,200	§	35	71	149	104	22.0
	400	750	76,200	58,200	36	71	149	104	22.0
	500	930	73,900	58,200	37	72	143	105	22.5
	600	1110	71,700	53,800	40	74	140	105	23.0
B	None		89,600	§	26	59	183	48	
	300	570	85,100	§	27	62	179	61	
	400	750	85,100	65,000	27	62	179	69	
	500	930	85,100	65,000	28	62	179	77	
	600	1110	80,600	60,500	32	66	170	88	

* Steel Research Committee.[55]

† Reported as yield point.

‡ Izod, 10 mm. (0.394 in.) sq. with 45-deg. notch, 0.25-mm. (0.098-in.) radius. Charpy, V-notch specimen of same dimensions as Izod. Each bar contained three specimens.

§ Could not be determined accurately.

Results obtained by the Steel Research Committee on water-quenched specimens containing 0.20 per cent carbon, 0.99 per cent manganese, 0.11 per cent silicon, 0.066 per cent sulphur, and 0.049 per cent phosphorus are interesting as they indicate the effect of tempering on this higher manganese material and thus afford, by comparison with the properties given in Fig. 54, an approximate evaluation of the effect of the manganese. The results are given in Table 48.

The difference between 0.99 per cent manganese and the usual manganese content of 1020 steel (0.30 to 0.60 per cent) is evidently responsible for an increase of approximately 10,000 lb. per sq. in. in tensile and yield strengths. The percentages of

TABLE 48.—EFFECT OF TEMPERING ON MECHANICAL PROPERTIES OF WATER-QUENCHED 0.20 PER CENT CARBON, 0.99 PER CENT MANGANESE STEEL*

Tempering temperature		Tensile strength, lb. per sq. in.	Yield strength,† lb. per sq. in.	Elongation in 2 in., per cent	Reduction of area, per cent	Brinell hardness	Izod impact, ft-lb.
°C.	°F.						
None		116,500	‡	18.0	46	241	33
300	570	107,500	‡	20.0	52	223	33
400	750	109,800	‡	20.0	52	...	45
500	930	103,000	78,400	23.0	57	217	61
600	1110	98,600	76,200	25.0	63	212	79

* Steel Research Committee.(55)
† Reported as yield point.
‡ Results by different laboratories not consistent.

elongation and reduction of area of the higher manganese material and the regular 1020 steel are practically the same.

Apparently the only data available on the effect of tempering quenched steels containing between 0.25 and 0.30 per cent carbon are those compiled by the Steel Research Committee on steels of the following analyses:

Element	Percentage	
	A	*B*
Carbon	0.26	0.26
Manganese	0.50	0.75
Silicon	0.20	0.30
Sulphur	0.060	0.044
Phosphorus	0.065	0.050
Nickel	0.03	0.15
Chromium	Nil	0.04

The essential difference in composition is the higher manganese in steel *B*, which also contained a small amount of nickel. The

TABLE 49.—EFFECT OF TEMPERING ON MECHANICAL PROPERTIES OF QUENCHED 0.26 PER CENT CARBON STEELS CONTAINING *A*—0.50 PER CENT MANGANESE AND 0.065 PER CENT PHOSPHORUS AND *B*—0.75 PER CENT MANGANESE AND 0.050 PER CENT PHOSPHORUS*

Steel	Tempering temperature		Quenched in water†						Quenched in oil†					
	°C.	°F.	Tensile strength, lb. per sq. in.	Yield strength,‡ lb. per sq. in.	Elongation in 2 in., per cent	Reduction of area, per cent	Brinell hardness	Izod impact, ft-lb.	Tensile strength, lb. per sq. in.	Yield strength,‡ lb. per sq. in.	Elongation in 2 in., per cent	Reduction of area, per cent	Brinell hardness	Izod impact, ft-lb.
A	As quenched		98,600	§	24	56	196	16	91,800	§	28	60	179	20
	300	570	96,300	§	24	57	192	14	91,800	§	28	60	179	20
	400	750	98,300	§	25	58	187	16	91,800	§	28	60	179	20
	500	930	94,100	71,700	27	60	183	19	89,600	65,000	29	61	174	22
	600	1110	89,600	65,000	29	64	179	22	87,400	60,500	31	63	171	24
	700	1290	82,900	58,200	33	68	167	26	80,600	56,000	34	68	156	27
B	As quenched		127,700	§	11	30	...	13	98,600	§	25	55	207	31
	300	570	127,700	§	12	32	...	13	98,600	§	26	57	207	39
	400	750	121,000	85,100	15	44	...	26	98,600	§	26	59	207	51
	500	930	107,500	80,600	23	58	...	56	96,300	69,400	27	60	202	65
	600	1110	96,300	69,400	29	64	...	76	89,600	62,700	29	64	187	79

* Steel Research Committee.(55)
† Steel *A* quenched from 870°C. (1600°F.); steel *B* from 900°C. (1650°F.).
‡ Reported as yield point.
§ Results by the different laboratories not consistent.

percentages of sulphur and phosphorus in these steels are higher than in normal commercial carbon steels.

The properties of these two steels after oil and water quenching and tempering at increasing temperatures are given in Table 49. The steel with the higher manganese had considerably higher tensile and yield strengths, especially after water quenching.

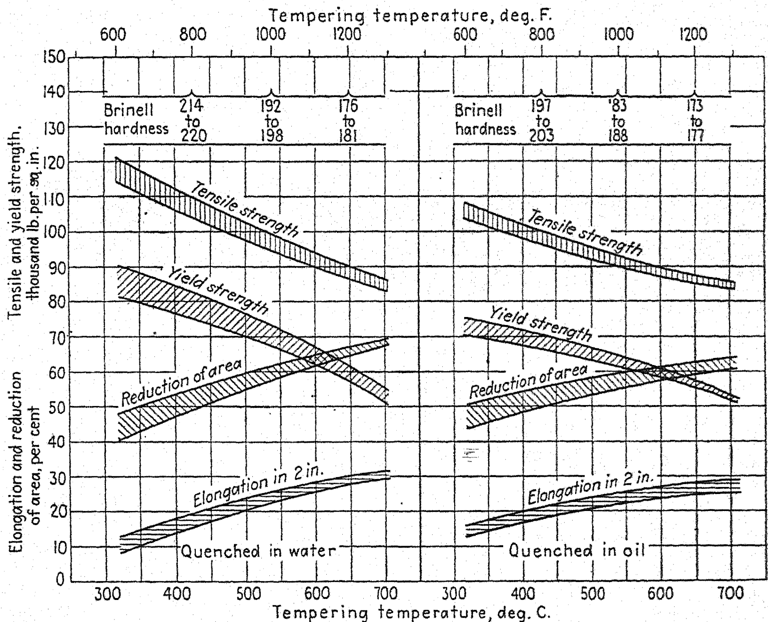

FIG. 55.—Effect of tempering on the tensile properties of 0.30 to 0.40 per cent carbon, 0.60 to 0.90 per cent manganese (1035) steel. Sections, 0.5 to 1.50 in. in diameter, were quenched in oil and water from 830 to 860°C. (1525 to 1575°F.) and tempered (about 60 min.) as shown.

The higher impact resistance of steel *B*, especially for tempering temperatures of 400°C. (750°F.) and above, is noteworthy.

76. Quenched and Tempered Structural and Forging Steels (0.31 to 0.60 Per Cent Carbon).—In contrast to the lower carbon grades, the use of heat-treated structural and forging steels containing 0.30 to 0.60 per cent carbon is fairly common; hence, there are many more data available on the effect of heat treatment on the properties of these higher carbon materials. For 0.30 to 0.40 per cent carbon (1035) steel, charts showing the effect of tempering after water and oil quenching have been published by the Society of Automotive Engineers[817] and French

and Sands.[700] In addition to these charts the U. S. Army Air Corps determined the effect of tempering on the properties of water-quenched 1035 steel, and Nead[35] investigated the properties of tempered oil-quenched material containing 0.32 per cent carbon (see page 84 for complete analysis). In every case the properties were determined on standard 0.505 × 2-in. test specimens, machined after heat treatment from bars 1.5 in. in

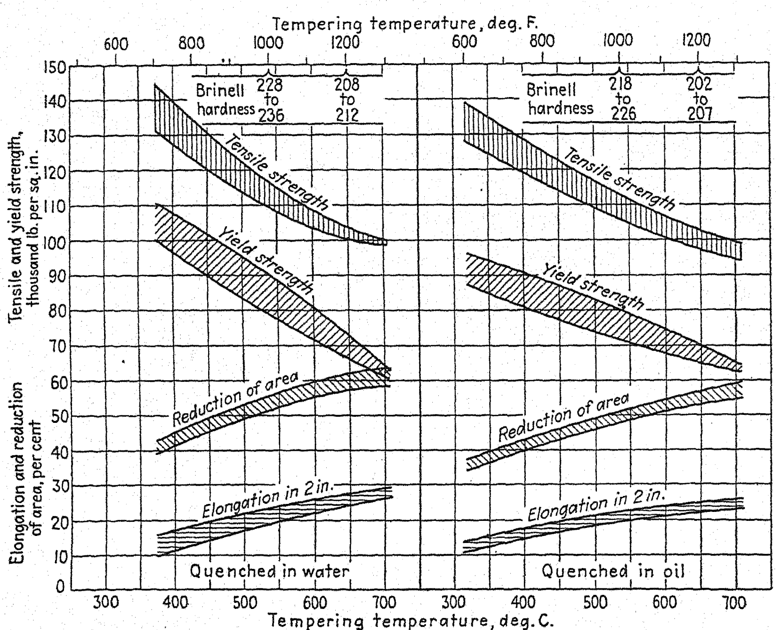

FIG. 56.—Effect of tempering on the tensile properties of 0.40 to 0.50 per cent carbon, 0.60 to 0.90 per cent manganese (1045) steel. Sections, 0.5 to 1.50 in. in diameter, were quenched in oil and water from 800 to 835°C. (1475 to 1525°F.) and tempered (about 60 min.) as shown.

diameter or smaller. The quenching temperature used by all of the investigators was 830 to 860°C. (1525 to 1575°F.).

All of these data have been used in preparing the curves shown in Fig. 55. In the charts for both water- and oil-quenched steels the values for tensile and yield strengths obtained by French and Sands are at the top or slightly above the upper limit of the hatched area and the values given on the S.A.E. standard property charts fall in the lower part. In values for elongation and reduction of area this order is reversed.

For the charts of 0.40 to 0.50 per cent carbon steel (1045) shown in Fig. 56, property curves by the Society of Automotive Engineers,[817] French and Sands,[700] and the U. S. Army Air Corps were used together with values obtained by Nead[35] on a 0.46 per cent carbon steel (see page 84 for complete analysis) and by the Steel Research Committee[55] on a steel containing 0.45 per

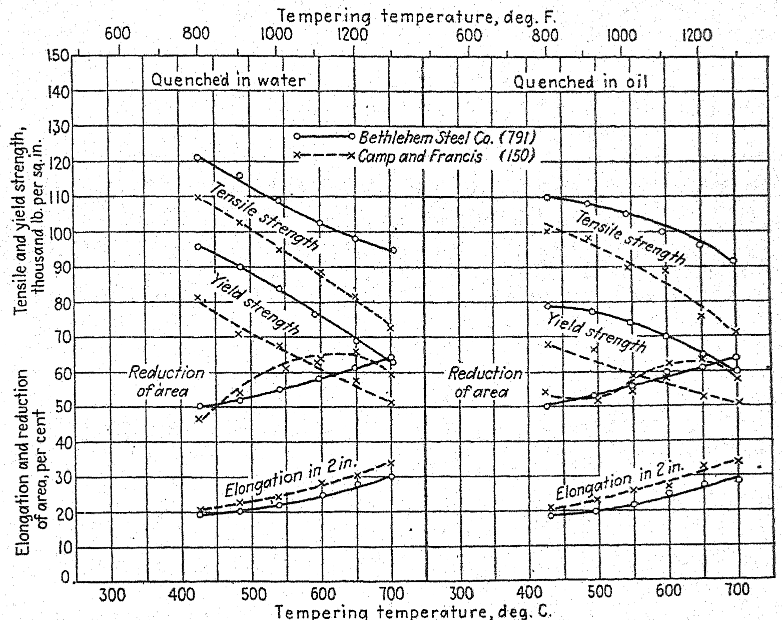

FIG. 57.—Effect of tempering on the tensile properties of 0.35 to 0.45 per cent carbon, 0.60 to 0.90 per cent manganese (1040) steel (*Bethlehem Steel Company*[791]) and of 0.38 per cent carbon, 0.55 per cent manganese steel (*Camp and Francis*[150]). Sections 1 in. in diameter were quenched in oil and water from 815 to 845°C. (1500 to 1550°F.) and tempered (about 60 min.) as shown.

cent carbon, 0.78 per cent manganese, 0.32 per cent silicon, and 0.025 per cent or less sulphur and phosphorus. Properties were determined by these investigators on standard test specimens machined from bars 1.5 in. in diameter or smaller and quenched from 800 to 830°C. (1475 to 1525°F.), except the specimens tested by the Steel Research Committee which were quenched from 870°C. (1600°F.).

To show the advantages of water quenching followed by tempering 0.30 to 0.50 per cent carbon steels, average mechanical properties from Figs. 55 and 56 have been tabulated against

uniform tensile strengths in Table 50. Thus it is clearly evident that, if 1035 steel is treated so that the tensile strength is 95,000 lb. per sq. in., slightly higher yield strength and markedly higher percentages of elongation and reduction of area will result from water quenching followed by tempering at the proper temperature to produce the required tensile strength than from oil quenching and tempering to produce the same tensile strength.

TABLE 50.—EFFECT OF WATER AND OIL QUENCHING ON THE YIELD STRENGTH, ELONGATION, AND REDUCTION OF AREA OF 0.30 TO 0.50 PER CENT CARBON STEELS OF THE SAME TENSILE STRENGTH

For a tensile strength of, lb. per sq. in.	Water quenched and tempered			Oil quenched and tempered		
	Yield strength, lb. per sq. in.	Elongation in 2 in., per cent	Reduction of area, per cent	Yield strength, lb. per sq. in.	Elongation in 2 in., per cent	Reduction of area, per cent
0.30 to 0.40 per cent carbon (1035)						
105,000	77,000	19	53	72,000	15	48
100,000	73,000	22	57	69,000	18	52
95,000	68,000	26	61	66,000	22	54
90,000	62,000	28	64	62,000	25	57
85,000	53,000	30	68	53,000	27	62
0.40 to 0.50 per cent carbon (1045)						
125,000	95,000	17.5	47	86,000	15.5	41
120,000	91,000	19	50	83,000	17	44
115,000	87,000	21	52	80,000	19	47
110,000	82,000	23	55	76,000	21	49
105,000	75,000	24	58	73,000	22	52
100,000	67,000	26.5	61	67,000	24	55

Data on the effect of tempering on the properties of quenched 0.35 to 0.45 per cent carbon (1040) steel are plotted in Fig. 57. The solid-line curves are from a standard property chart[791] and fall near the lower boundary for tensile and yield strengths of the hatched area given in Fig. 56 for 1045 steel. These curves are considerably above the corresponding areas in Fig. 55 for 1035 steel.

Camp and Francis[150] determined the properties of standard test specimens machined from a 1-in. round bar which had been rolled from a billet containing 0.38 per cent carbon, 0.55 per cent manganese, 0.05 per cent silicon, 0.050 per cent sulphur, and 0.024 per cent phosphorus. The quenching temperature was 815°C. (1500°F.).

Using the data available from Figs. 55, 56, and 57, it should be possible to average the values so that a chart for a steel containing

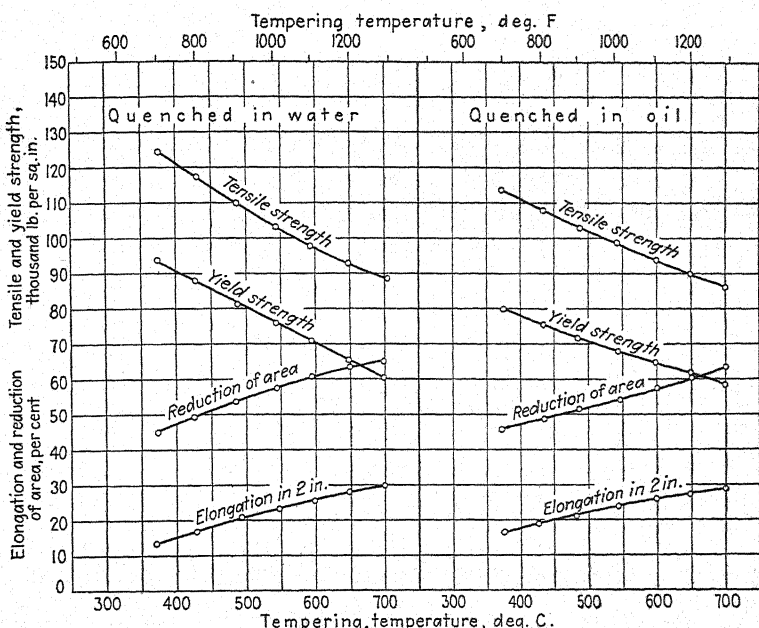

FIG. 58.—Most-probable tensile properties, for sections of 0.5 to 1.5 in. in diameter, of water-quenched and tempered and oil-quenched and tempered steels containing 0.37 to 0.43 per cent carbon and 0.60 to 0.90 per cent manganese.

approximately 0.37 to 0.43 per cent carbon could be constructed which should indicate with considerable accuracy the mechanical properties which would result most frequently from each tempering treatment. This has been attempted in Fig. 58. The curves in this chart, like the curves in Figs. 55 and 56, indicate that for the same tensile strength, water quenching followed by suitable tempering will result in higher yield strength, elongation, and reduction of area than oil quenching and tempering to produce the same tensile strength.

Apparently the only data available on steels containing between 0.50 and 0.60 per cent carbon are those of Nead,[35] who quenched a steel containing 0.57 per cent carbon in oil from 810°C. (1490°F.) and tempered at increasing temperatures. The complete composition is given on page 84. Nead's bars were slightly larger than the standard 0.505 × 2-in. specimens and were ground or machined to size after heat treatment. The properties are given in Table 51.

TABLE 51.—EFFECT OF TEMPERING ON THE MECHANICAL PROPERTIES OF A STEEL CONTAINING 0.57 PER CENT CARBON AND 0.65 PER CENT MANGANESE, OIL QUENCHED FROM 810°C. (1490°F.)*

Tempering temperature		Tensile strength, lb. per sq. in.	Yield strength,† lb. per sq. in.	Elongation in 2 in., per cent	Reduction of area, per cent	Brinell hardness
°C.	°F.					
None		152,000	105,000	16.5	40.3	311
370	700	153,000	106,500	16.0	43.3	302
460	860	145,500	97,500	16.0	46.2	293
560	1040	133,000	93,500	20.5	51.9	255
650	1200	113,500	79,500	24.0	62.3	228

* Nead.[35]

† Reported as yield point.

D. QUENCHED AND TEMPERED HIGH-CARBON STEELS

With the exception of rails and a few other railway materials such as wheels and tires, nearly all carbon steels containing more than 0.60 per cent carbon are used commercially in the heat-treated condition. A large part of this production is made into tools, dies, and springs. The chief requirement for tools is hardness; for dies and springs, hardness and a certain amount of toughness are necessary.

Depending upon the kind of tool and the service required from the material, high-carbon tool steels contain between 0.65 and 1.25 per cent carbon. They are usually quenched in cold water and tempered at 150 to 340°C. (300 to 650°F.): at the lower temperatures of this range when it is only necessary to relieve strains, and at the higher temperatures to increase the toughness. Carbon steels for dies and springs contain 0.50 to 1.05 per cent carbon; die steels are water quenched, springs are usually oil

quenched; both are tempered to give the hardness and toughness required. The heat treatments for these various high-carbon materials have been carefully standardized[821] and will not need to be discussed further.

The effect of quenching and tempering on the structure and physical properties of high-carbon steels has been discussed in detail in Volume I of this monograph (Chapters VI, VII, and IX). It will, therefore, suffice in this chapter to give attention only to the effect of tempering on the tensile properties.

77. The Brittleness of Quenched and Tempered High-carbon Steel.—The tensile properties of high-carbon steel are of minor practical importance as compared with the hardness; hence, there are few data available. Moreover, many of these data, as will be shown hereafter, are of little value. It is well known that fully hardened high-carbon steel is very brittle, and that quenching produces severe internal stresses which may cause immediate or later cracking and even complete disintegration of a tool or other finished section. This is discussed in detail in Volume I of this monograph (pages 316 to 329). In many cases, internal stresses may be relieved and cracking prevented by tempering. In other cases, however, internal cracks may form during or just after quenching which cannot be healed by ordinary reheating.

Rawdon and Epstein[93] showed in 1922 that microscopic cracks may form in etched martensite, and Lucas[469] pointed out in 1931 that these cracks were evident after etching even when the steel had been tempered as high as 205°C. (400°F.). Davenport, Roff, and Bain[684] recently found that the brittleness of fully hardened and tempered steel is due, in part at least, to the presence of real or potential microscopic quenching cracks, and that these cracks, or their causes, may not be wholly removed by tempering once they are developed in the martensite.

These investigators treated specimens of a steel containing 0.74 per cent carbon, 0.37 per cent manganese, 0.145 per cent silicon, 0.039 per cent sulphur, and 0.044 per cent phosphorus so that the specimens differed only with respect to the cracks; composition, grain size, quenching temperature, and final hardness were identical. Specimens, 0.180 in. in diameter, were heated to 790°C. (1450°F.), held 5 min., and quenched into a lead-alloy bath at 305°C. (580°F.). After 15 min. in the lead

bath, the specimens were quenched in water. Other specimens of the same size were heated to the same temperature for the same time and quenched into oil at 20°C. (70°F.) and tempered immediately in a lead-alloy bath at 315°C. (600°F.). They were held at this temperature for 30 min. and also quenched into water. The resulting properties were as follows:

Property	Quenched	
	In lead	In oil
Rockwell *C* hardness	50.4	50.2
Tensile strength, lb. per sq. in.	282,700	246,700
Yield strength (yield point), lb. per sq. in.	151,300	121,700
Elongation in 6 in., per cent	1.9	0.3
Reduction of area, per cent	34.5	0.7
Impact on 0.180-in. round, unnotched specimen, ft-lb	35.3	2.9

Producing a crack-free steel by quenching in a lead-alloy bath at 305°C. (580°F.) and holding at this temperature until the austenite was completely transformed was responsible for marked improvement in tensile and yield strengths and also in elongation, reduction of area, and impact resistance.

Davenport, Roff, and Bain also found that

> the severity of the microcracking, and the consequent lowering of ductility, increases with the grain size of the austenite from which the tempered martensite is developed in the heat treating. Large austenite grain size has an injurious effect upon the ductility even of crack-free hardened steel, but the effect of microcracks is large in comparison with the influence of grain size alone.*

Hughes and Dowdell[721] compared the mechanical properties and structure of 0.5-in. specimens of 0.34 and 0.49 per cent carbon steels quenched into hot lead with the properties and structure of the same material quenched in oil or water and tempered. No cracks were found in any of the specimens, but tempering temperatures as high as 480 to 650°C. (900 to 1200°F.) were necessary to produce the same tensile properties and structure

* The relation between grain size and mechanical properties is discussed in the next chapter.

in the oil- or water-quenched specimens as were obtained by quenching from the same temperature into lead maintained at 345°C. (650°F.).

78. Properties of Quenched and Tempered High-carbon Steels.—Because of the likelihood of the formation of microcracks in water or oil quenching high-carbon steel, and because of the effect of the austenite grain size, neither of which was taken into consideration by investigators until recently, there is reason to suspect the accuracy of nearly all of the older work on the properties of quenched and tempered high-carbon steel. Thus, the comprehensive investigation by Brinell as reported by Wahlberg(5) in 1901 on tempering water- and oil-quenched high-carbon steels may be dismissed with the bare statement that specimens were quenched in oil and water from 750, 850, and 1000°C. (1380, 1560, and 1830°F.), and mechanical properties were determined on specimens as quenched and as tempered at 350, 550, and 650°C. (660, 1020, and 1200°F.). The principal conclusion to be drawn from Brinell's data is that, in general, the use of higher quenching temperatures resulted in higher tensile and yield strengths and lower elongation and reduction of area.

Three sets of data on the effect of tempering quenched high-carbon steel will be given, although it is possible that microscopic cracks of which the original investigators were unaware have affected the results to an unknown degree. Nead(35) quenched oversize 0.505 × 2-in. gage specimens of high-carbon steels in oil (see pages 84 and 178 for complete analyses and quenching temperatures) and tempered them at 375, 460, 560, and 650°C. (705, 860, 1040, and 1200°F.) for 30 min. After grinding to size, the mechanical properties were determined.

Nead's data on the tensile strength of the specimens containing 1.01 and 1.46 per cent carbon tempered at the lower temperatures, on the yield strength of the specimens containing 0.83 and 1.46 per cent carbon tempered at the lower temperatures, and on reduction of area of the specimens containing 0.71 and 1.22 per cent carbon tempered at the lower temperatures and of the 1.46 per cent carbon specimen tempered at all temperatures were so improbable that they were not reproduced. Also, all of the data on the specimens containing 1.39 per cent carbon were considered to be too far from the truth to be reproduced. Moreover, it is very probable that some of the points given in Fig. 59,

for example the tensile strength of specimens containing 0.71, 0.83, and 1.22 per cent carbon tempered at 375°C. (705°F.), were affected by the presence of internal stresses or microcracks. This applies also to some of the values for reduction of area. Regardless of the absolute accuracy of Nead's data, they are of value in showing that tempering reduces the tensile and yield strengths in a fairly uniform manner as the temperature is increased.

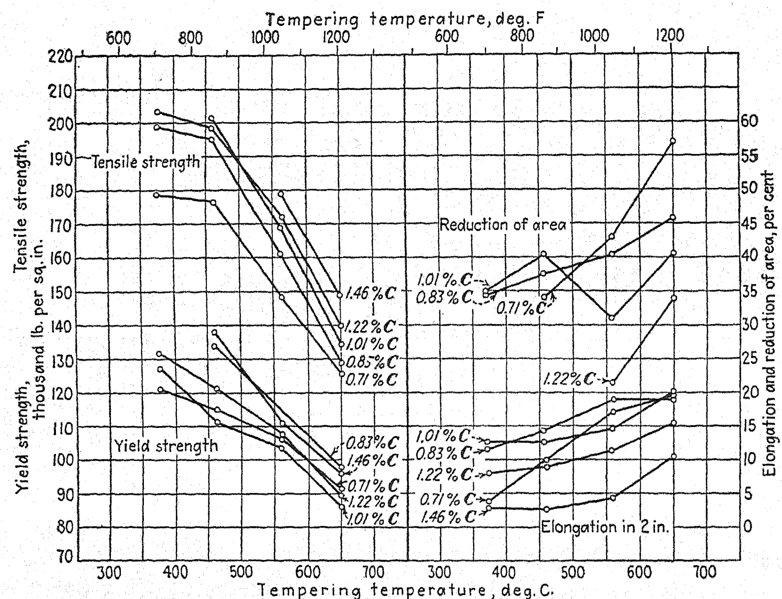

Fig. 59.—Effect of tempering on the tensile properties of oil-quenched 0.505-in. specimens of high-carbon steel. (*Nead.*[35])

French and Johnson[84] determined the effect of tempering on the mechanical properties of oil- and water-quenched steel containing 1.04 per cent carbon, 0.17 per cent manganese, 0.14 per cent silicon, and less than 0.020 per cent sulphur and phosphorus. Specimens approximately 0.5 × 2 in., machined from 1-in. round bars which had been normalized at 815°C. (1500°F.), were quenched in oil and in water after holding 30 min. at 790°C. (1450°F.). They were tempered 30 min. at the temperatures shown by the points in Fig. 60. From their results French and Johnson concluded that 1 per cent carbon steels should be quenched from just above the Ac_1 temperature and that for

the higher tempering temperatures water quenching is to be preferred as it results in a much higher elastic ratio (Fig. 60).

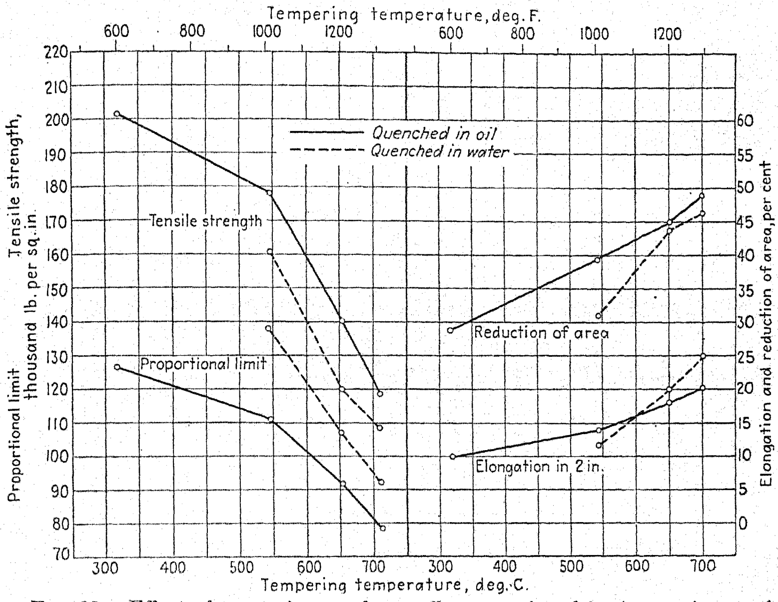

FIG. 60.—Effect of tempering on the tensile properties of 0.5-in. specimens of 1.04 per cent carbon, 0.17 per cent manganese steel quenched in oil and water from 790°C. (1450°F.) and tempered 30 min. at indicated temperatures. (*French and Johnson.*[84])

French and Johnson also investigated the effect of increasing the quenching temperature.

For bars tempered at 540°C. (1000°F.), the strength, proportional limit, and hardness of oil-quenched specimens increase and maximum values are obtained after quenching from 845°C. (1550°F.) which is coincident with retention of all but a small part of the excess cementite. A higher quenching temperature results in decreased strength.

Quenching in oil from 845°C. (1550°F.) resulted in lower values for elongation and reduction of area than those plotted in Fig. 60 for a quenching temperature of 790°C. (1450°F.).

Standard property charts given by the Bethlehem Steel Company,[791] showing the effect of tempering on water- and oil-quenched high-carbon steels, are reproduced in Fig. 61. The composition ranges and quenching temperatures of the materials are as follows:

Element	Percentage	
	1080	1095
Carbon	0.75 to 0.90	0.90 to 1.05
Manganese	0.30 to 0.50	0.25 to 0.50
Silicon	0.15 to 0.25	
Sulphur	0.05 max.	0.055 max.
Phosphorus	0.04 max.	0.04 max.
Quenched in oil from	815°C. (1500°F.)	800°C. (1475°F.)
Quenched in water from	800°C. (1475°F.)	790°C. (1450°F.)

Specimens were 0.505 × 2 in. and were machined (or ground) from 1-in.-diameter heat-treated bars. Before quenching, the bars were normalized at 900°C. (1650°F.). Properties were

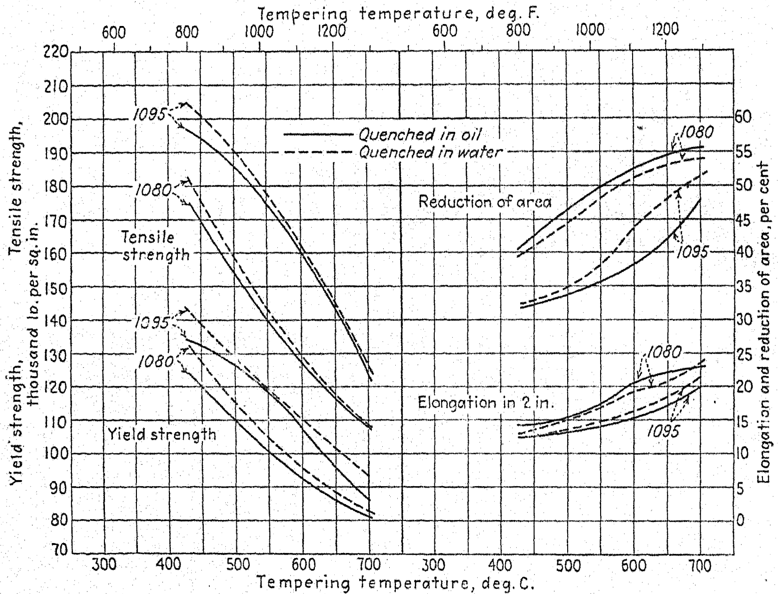

FIG. 61.—Effect of tempering on the tensile properties of oil- and water-quenched high-carbon steel. Composition and heat treatment are given above. (*Bethlehem Steel Company.*[791])

determined after tempering at 425, 540, and 650°C. (800, 1000, and 1200°F.). There are no details how many values were determined for each tempering temperature or how closely the individual determinations checked.

If the curves for the properties of oil-quenched and tempered 1.01 per cent carbon steel (Nead), 1.04 per cent carbon steel (French and Johnson), and 1095 steel (Bethlehem Steel Company) are compared, it will be noted that tensile-strength and elongation values check closely. French and Johnson's values for proportional limit are slightly higher than Nead's values for yield strength (yield point), and the Bethlehem Steel Company's values for yield strength are 10,000 to 15,000 lb. per sq. in. higher than the values of the other investigators. The reduction of area values by French and Johnson and by the Bethlehem Steel Company check closely. There is a wide variation between tensile- and yield-strength values for the water-quenched 1.04 per cent carbon steel as determined by French and Johnson and 1095 steel as determined by the Bethlehem Steel Company for all tempering temperatures. Values for elongation check closely and for reduction of area fairly well.

As the effect of tempering on the hardness of high-carbon steel is discussed in detail in Volume I of this monograph (pages 221 to 226), it is unnecessary to reproduce the data here. The following is a brief summary of Epstein's discussion.

1. The hardness of quenched carbon steels tends to decrease gradually on tempering, the hardness versus tempering-temperature curve being fairly smooth except for a rather sharp change in slope downward beginning at about 250°C. (480°F.).

2. The time of tempering has an appreciable effect on the resulting hardness, although the magnitude of the time factor, for the lower tempering temperatures, is not so great as might be supposed. At relatively low tempering temperatures about 75 per cent of the total softening which occurs in tempering for 25 hr. will take place in 20 min., and at higher temperatures in a much shorter time. If a small section of quenched eutectoid steel has a scleroscope hardness of 62 after tempering 15 min. at 500°C. (930°F.), it will have a scleroscope hardness of 58 after tempering 45 min. and of 56 after tempering 2 hr. and 15 min. at the same temperature.

E. AUTHOR'S SUMMARY

1. It might be tentatively concluded, subject to modification when more data become available, that (*a*) normalizing commercial hot-worked carbon steels improves the elongation and reduc-

tion of area and has little or no effect on the tensile and yield strengths; and (*b*) normalized carbon steels of fairly high purity (deoxidized with aluminum) have markedly lower tensile and yield strengths and, perhaps, in the medium-carbon ranges higher elongation and reduction of area than normalized commercial carbon steels.

2. Available data on the properties of annealed low- and medium-carbon steels are conflicting (see Figs. 48 and 49).

3. Results quoted (section 70) show that tensile strength, yield strength, and hardness decrease and elongation, reduction of area, and impact resistance increase when high-carbon steels are annealed or normalized in the temperature range 700 to 850°C. (1290 to 1560°F.); these properties return to their original values when the steels are annealed or normalized at higher temperatures.

4. In annealing steels containing 0.74 and 0.94 per cent carbon, there is little or no change in properties if the time for which the material is held at the annealing temperature is increased from 30 min. to 5 hr.

5. The conclusion given in No. 1 above, that the tensile strength of hot-worked steel is affected little if at all by normalizing, is valid only if the normalizing temperature is 900°C. (1650°F.) or higher, and possibly only if the carbon is less than 1 per cent.

6. Ingot iron (0.01 to 0.03 per cent carbon) and very low carbon steel can be hardened by quenching. In the former, water quenching increases the tensile strength about 15 per cent, the yield strength about 65 per cent, and the hardness about 35 per cent. The elongation is reduced 20 per cent and the reduction of area is unchanged. Quenching 0.05 per cent carbon steel drastically in water doubles the tensile strength and halves the elongation. Water quenching a 0.15 per cent carbon rivet steel increases tensile strength about 30 per cent and yield strength about 65 per cent, as compared with the properties of the same material after annealing. The elongation is lowered about 30 per cent.

7. The tensile strength, yield strength, and Brinell hardness of untempered small sections of oil-quenched steel increase linearly with the carbon content up to approximately 0.7 per cent. The elongation, reduction of area, and impact values decrease with increasing carbon content. Above 0.7 per cent carbon,

internal stresses and possible microscopic cracks produced by quenching may affect greatly the properties of quenched specimens; hence, published data for these higher carbon materials may not represent true properties of unstressed or uncracked material.

8. Brinell-hardness values indicate that, except in very small sections of high-carbon steel, full hardening does not result from oil quenching. Water quenching medium- or high-carbon steels apparently produces full hardening in small sections but results in such severe internal stresses and extreme brittleness that published data on mechanical properties of untempered water-quenched steel containing more than 0.20 to 0.30 per cent carbon have little or no meaning.

9. Low-carbon steels (less than 0.30 per cent carbon) are seldom used commercially in the quenched and tempered condition, hence, data on their properties are relatively scarce. Values are given in sections 74 and 75 which show the effect of tempering on the properties of quenched ingot iron, 0.05 per cent carbon, 0.12 per cent carbon, 0.20 per cent carbon, and 0.26 per cent carbon steels, some of which have relatively high manganese.

10. Property charts showing the effect of tempering are common in the literature but are subject to many errors through variations in composition and variables introduced by differences in manufacture, grain size, heat-treating practice, and testing technique. Despite these sources of error, these charts are of value in indicating the properties which may be expected *most frequently;* hence, new charts have been prepared for steels containing 0.15 to 0.25 per cent carbon (1020), 0.30 to 0.40 per cent carbon (1035), 0.40 to 0.50 per cent carbon (1045), and, from a weighted average of the last two and other data, 0.37 to 0.43 per cent carbon (1040) steel. In all charts but the one for 1040 steel the expected properties have been indicated by an area instead of a line.

11. These property charts and other data indicate that the best combination of properties for medium-carbon (0.30 to 0.50 per cent) steels results from water quenching and tempering. Thus, for equal tensile strength, steels which have been water quenched and tempered have slightly higher yield strength and in some cases markedly higher elongation and reduction of area.

12. Limited data indicate that increasing the manganese content of low- and medium-carbon steels from about 0.50 to

about 0.80 per cent increases the tensile and yield strengths 10,000 lb. per sq. in. or more, depending upon the heat treatment, and lowers the elongation and reduction of area from a very small amount to as much as 30 per cent.

13. Data are presented to show that in an oil-quenched 0.74 per cent carbon steel, even when tempered at 305°C. (580°F.), microcracks may be present which affect the properties deleteriously. A crack-free steel of the same material treated to the same degree of hardness has markedly higher tensile and yield strengths, and very much higher elongation, reduction of area, and impact resistance.

14. A large part of the older data on the properties of quenched and tempered high-carbon steel is of doubtful accuracy because of the probability of internal stresses and the possibility of microscopic cracks which would affect the results and which were not considered by the early investigators. Available concordant data on the properties of oil-quenched and tempered steel containing approximately 1 per cent carbon indicate that tempering has the following effect:

Tempering temperature		Tensile strength, lb. per sq. in.	Yield strength, lb. per sq. in.	Elongation in 2 in., per cent	Reduction of area, per cent
°C.	°F.				
425	800	197,000	120,000	12	33
540	1000	178,000	110,000	14	38
650	1200	140,000	92,000	18	44
700	1300	118,000	80,000	21	48

15. From the small amount of data available, it may be concluded that, for the higher tempering temperatures, water quenching a 1.00 to 1.10 per cent carbon steel produces a higher elastic ratio than oil quenching. If a reasonable amount of ductility, as measured by elongation and reduction of area, is required, quenched high-carbon steel should be tempered at 540 to 650°C. (1000 to 1200°F.). For the high hardness desired in tools and dies, high-carbon steels are quenched in water and tempered at 150 to 340°C. (300 to 650°F.). The effect of such tempering on the hardness has been discussed in Volume I, Chapter VII, of this monograph.

CHAPTER VII

EFFECT OF CROSS-SECTION AND OTHER VARIABLES ON MECHANICAL PROPERTIES

Effect of Cross-section on Mechanical Properties of Low-carbon Steels—Effect of Cross-section on Mechanical Properties of Medium-carbon Steels—Effect of Cross-section on Mechanical Properties of High-carbon Steels—Effect of Grain Size and Other Variables—Author's Summary

The amount of static stress which may be imposed upon a piece of wrought carbon steel before failure occurs depends upon a large number of variables. The three most important of these variables have been discussed in the previous chapters. For commercial carbon steels these are: (1) the amount of carbon, (2) the mechanical treatment, *i.e.*, whether the material was hot or cold worked and the variations in hot- and cold-working practice, and (3) the thermal treatment, if any, which the material underwent during or subsequent to mechanical treatment.

There are also other variables, probably less important than these three, which are discussed in the present chapter. These are: (*a*) the effect of variations in cross-section, the so-called mass effect, (*b*) the effect of grain size, and (*c*) the effect of aging or resting for relatively long periods of time at or near room temperature on the properties of high-carbon steels. Of these, the mass effect is probably the most important, although enthusiasts on grain-size control may disagree. Aging effects in high-carbon steel are of little practical interest and have, therefore, received only brief mention. All three of the variables discussed in the present chapter are of importance only—or at least of importance primarily—if the steel has been subjected to a prior heat treatment.

A. EFFECT OF CROSS-SECTION ON MECHANICAL PROPERTIES OF LOW-CARBON STEELS

As was emphasized in Volume I of this monograph, the hardness and strength of a quenched iron-carbon alloy containing

more than a very small amount of carbon depend primarily upon the rate with which the heat is extracted in quenching. As the quenching medium acts very rapidly upon the surface of the piece, this surface area and the layers near the surface cool quickly. In very small sections the rate of cooling at the center is controlled chiefly by the properties of the quenching medium, *i.e.*, by the rate at which the coolant can abstract heat from the surface. As the cross-section increases, however, the rate at which heat can pass through the steel from center to surface—the diffusivity—becomes of increasing importance, and in large masses it is the controlling factor.

79. Importance of the Mass Effect.—In sections of large mass the cooling rate on quenching decreases as the distance from the surface increases, and, as a corollary to this, hardness and strength decrease progressively from the surface to the center. The greatest difference in properties is found in untempered material; this difference diminishes progressively as the steel is tempered at increasing temperatures. For example, assume that a 2-in. section as quenched has a Brinell hardness of 500 at the surface and 300 at the center. After tempering at 400°C. (750°F.) the Brinell-hardness values will be 400 and 300 respectively; after tempering at 600°C. (1110°F.) the hardness will be the same throughout the cross-section, 300 at center and surface.

The importance of mass effect is clear. For most of the commercial applications of heat-treated carbon steel the engineer must know the strength and hardness of the entire section which is being used; to him the average properties, or in some cases the minimum properties, are the important ones, not the properties of the surface layers. It is, however, usually impracticable to determine the average properties by testing an entire large section (for example an automobile drive shaft or a locomotive drive rod), consequently, in most cases, the strength must be ascertained from small specimens which can be broken in the usual laboratory machine.

As the change in properties with mass is not usually a linear function of the distance from the surface, a large number of investigations have been made to determine the actual mechanical properties in heat-treated carbon and alloy steels of different cross-sections, some of them under the direction of such organizations as the Society of Automotive Engineers in the United States

and the Steel Research Committee of the Institution of Automobile Engineers of Great Britain. Some of these data are summarized hereafter.

80. Low-carbon Steels of Moderate Cross-section.—The most extensive set of data on the effect of section on the mechanical properties of carbon and alloy steels, reported in 1920, was obtained by the Steel Research Committee[55] mentioned above. These data are especially important as they were secured on steels of definite composition and not on material having a composition range. The Committee treated hot-rolled rods under carefully controlled conditions. British standard specimens (0.564 in. in diameter with a 2-in. gage) were cut from the core of each rod. Tests were made by three different laboratories and the results correlated, weighted, and averaged. With the exception of some of the yield-strength values, the results obtained by the different laboratories were in good agreement.

The effect of increasing the section from $^{11}/_{16}$ to $2\frac{1}{8}$ in. on the tensile and impact properties of two low-carbon steels is plotted in Fig. 62. In both of these steels the sulphur and phosphorus were relatively high: for *A* the percentages were 0.080 and 0.048, and for *B* 0.066 and 0.049 respectively. The double quenching in water was used to simulate the treatment frequently given to carburized material: namely, water quenching from 900°C. (1650°F.) to refine the core, followed by water quenching from 760°C. (1400°F.) to harden the case.

As would be expected, the tensile- and yield-strength values decrease with increase in section—the decrease being greater with the higher carbon steel—and the elongation and reduction-of-area values increase. The peak in Izod impact values, in both steels, for the specimens of 1 to 1.5-in. diameter is noteworthy.

The properties of a steel containing 0.21 per cent carbon and 0.64 per cent manganese were determined by Lee and Schade[387] after quenching bars, ¼ and ½ in. thick, in water. As quenched, the ¼-in. section had a tensile strength of 142,000 lb. per sq. in. and the ½-in. section a strength of 116,500 lb. per sq. in.; after tempering at 480°C. (900°F.) the tensile strength of both sections was 99,000 lb. per sq. in.

81. Heavy Sections of Low-carbon Steel.—Maurer and Korschan[636] recently reported the results of a study of the properties of heat-treated 100-ton forgings of carbon, manganese,

nickel, and nickel-chromium-molybdenum steels. The average analysis of the carbon steel was 0.17 per cent carbon, 0.53 per cent manganese, and 0.27 per cent silicon; it was a mixture of acid and basic open-hearth materials and was cast into an octagonal ingot whose inscribed diameter at the base was 2030 mm. (80 in.) and whose length was approximately 3500 mm. (137.8 in.). The ingot was hot pressed to give three abutting cylinders

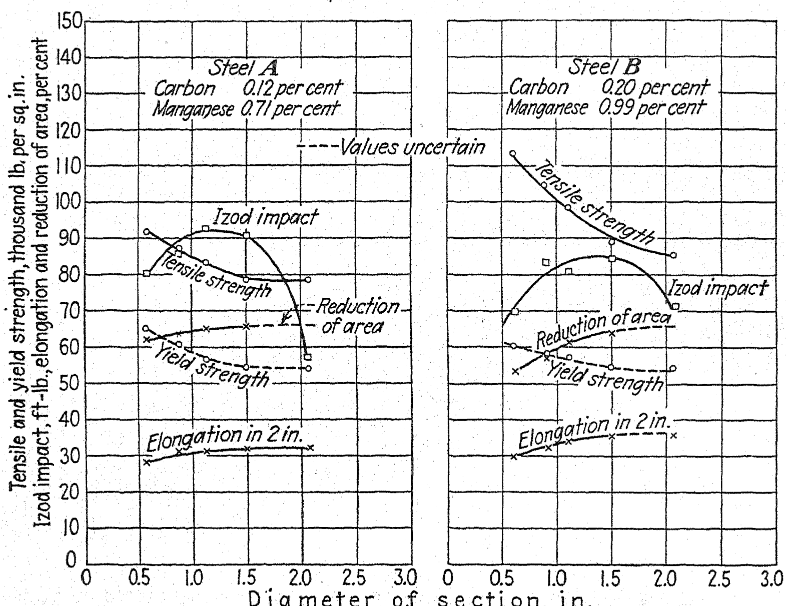

FIG. 62.—Effect of section size on the mechanical properties of low-carbon steel quenched in water from 900°C. (1650°F.), reheated to 760°C. (1400°F.), and again quenched in water. Specimens tested were 0.564 × 2 in., machined from the core of round bars heat treated in the sizes indicated. (*Steel Research Committee.*[55])

having diameters of 1450, 1180, and 920 mm. (57.1, 46.5, and 36.2 in.), which represent elongations of 2, 3, and 5 times respectively or reductions in cross-sectional area of 50, 67, and 80 per cent. After forging, the three cylindrical sections were cut apart and annealed at a temperature between 620 and 630°C. (1150 and 1165°F.). The cylinders were then cooled in air from 900°C. (1650°F.) and tempered at 620 to 630°C. (1150 to 1165°F.). A hole, the diameter of which was approximately one-tenth of the diameter of the cylinder, was then drilled through the axis of each cylinder, and the cylinders were again heat

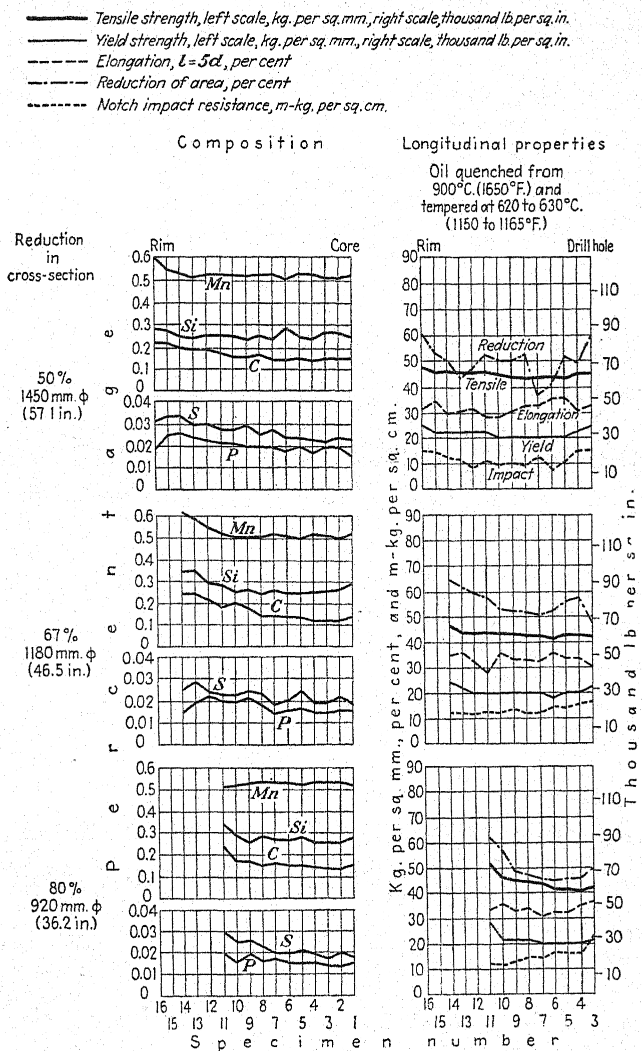

FIG. 63.—Composition and longitudinal properties throughout the cross-section of large forgings with an average composition of 0.17 per cent carbon, 0.53 per cent manganese, and 0.27 per cent silicon. Each specimen number represents a section of about 1.5 in., beginning at the core of the forging. (*Maurer and Korschan.*[636])

treated by quenching in oil from 900°C. (1650°F.) and tempering at 620 to 630°C. (1150 to 1165°F.). After each treatment,

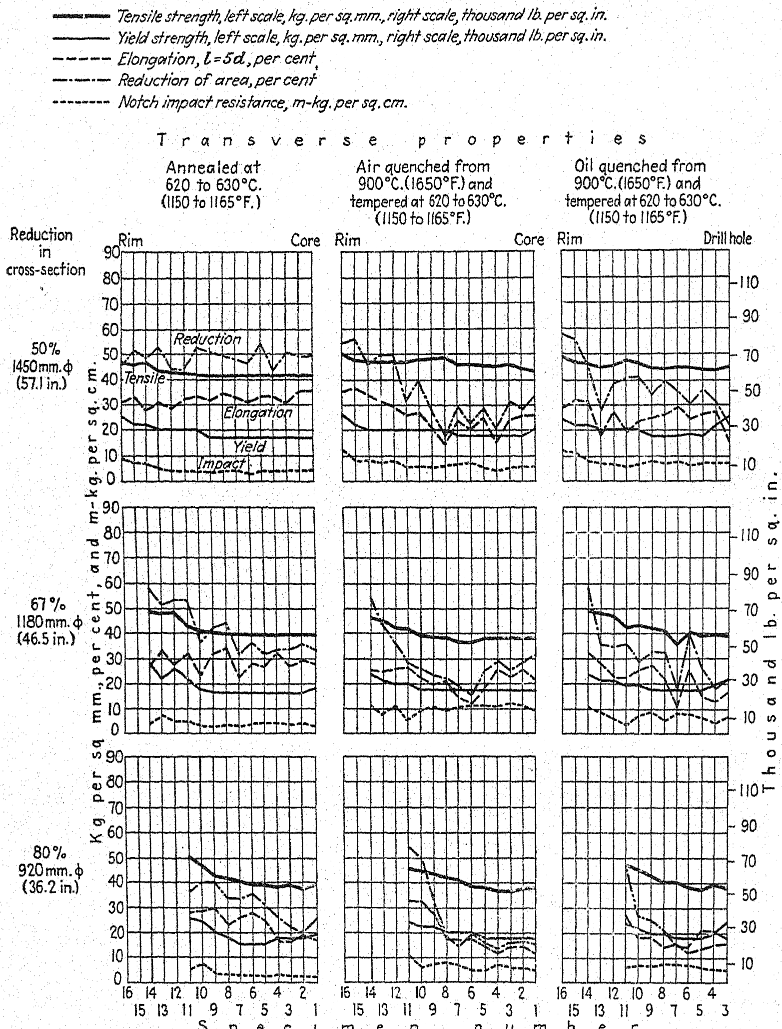

FIG. 64.—Transverse properties of large forgings whose composition and longitudinal properties are given in Fig. 63. (*Maurer and Korschan.*[f36])

sections for testing were cut from each cylinder. Tensile and impact specimens were cut at different locations between the surface and the center of each section. Specimens whose axes

were at right angles to the direction of elongation in pressing (tangential specimens) were considered to give the most useful measure of the properties of the forgings, and the properties in a direction parallel to the elongation were reported only for the oil-quenched and tempered material.

TABLE 52.—AVERAGE PROPERTIES THROUGHOUT THE CROSS-SECTION OF A VERY LARGE FORGING CONTAINING 0.17 PER CENT CARBON, 0.53 PER CENT MANGANESE, AND 0.27 PER CENT SILICON*

Diameter, mm.	Diameter, in.	Treatment†	Direction‡	Tensile strength, lb. per sq. in.	Yield strength, lb. per sq. in.	Elongation ($l = 5d$), per cent	Reduction of area, per cent	Charpy impact, m-kg. per sq. cm.
920	36.2	None	T	57,300	25,900	20.2	23.0	3.9
		Annealed	T	59,200	26,500	23.6	31.0	3.8
		Air quenched	T	57,100	28,200	18.5	24.4	6.6
		Oil quenched	T	59,000	29,000	17.4	24.0	6.6
		Oil quenched	L	63,100	31,000	33.6	49.6	15.4
1180	46.5	Annealed	T	58,900	26,600	28.5	40.7	3.8
		Air quenched	T	57,200	27,000	22.0	29.3	10.1
		Oil quenched	T	60,600	28,500	22.3	32.4	7.1
		Oil quenched	L	61,100	29,000	33.1	55.6	13.2
1450	57.1	Annealed	T	61,000	27,800	31.9	48.6	3.9
		Air quenched	T	65,700	28,000	26.0	34.0	6.1
		Oil quenched	T	65,200	29,300	25.4	39.0	7.9
		Oil quenched	L	63,700	30,600	31.4	50.0	11.0

* Maurer and Korschan.[636]
† See top of Figs. 63 and 64 for quenching and tempering temperatures.
‡ L = longitudinal, T = transverse.

The data obtained by Maurer and Korschan on the carbon steel are summarized in Figs. 63 and 64. The curves at the left of Fig. 63 show how the composition of the three sections varied from surface to core (the specimens were numbered beginning at the core). Table 52 gives the average properties of the material of the three sections after the different heat treatments. In general, there is a tendency for the strength to decrease from surface to core, which decrease may be accompanied by a decrease in ductility. Oil or air quenching followed by tempering improved the impact resistance even of the cylinder 1450 mm. (57.1 in.) in diameter. Longitudinal specimens, of the oil-quenched material, yielded higher values for elongation and

reduction of area and much higher values for Charpy impact resistance than transverse specimens, but the transverse properties are not nearly so poor as would be expected on the basis of Charpy's work mentioned in Chapter IV.

All the data which have been discussed so far in this chapter have been obtained from hot-worked and heat-treated steels. The size of the section has an appreciable effect also on the mechanical properties of cast steels. This was discussed in detail in Chapter II (pages 43 to 49) where it was shown that in general better mechanical properties can be secured in castings of small section than in large ones, and that in the latter the core of the casting is inferior in properties to material near the surface. When castings are heat treated, the effects of mass are much the same as when a hot-rolled section is heat treated; namely, that tensile and yield strengths are comparatively high in the portions near the surface and decrease as the distance from the surface increases.

B. EFFECT OF CROSS-SECTION ON MECHANICAL PROPERTIES OF MEDIUM-CARBON STEELS

There are many more data available on the effect of mass on the mechanical properties of medium-carbon steels than of low-carbon steels. This is natural; most of the low-carbon steels are not heat treated, and, although this class of material is used extensively in the form of heavy sections (structural steel, for instance), the slight difference in the mechanical properties throughout the section of a hot-rolled piece is usually not of major importance. The medium-carbon forging steels, on the contrary, are widely used in large heat-treated sections for highly stressed structural members; consequently, the effect of mass becomes of primary importance. This is the reason why most of the published data on the properties of carbon steels of different cross-sections have been obtained on S.A.E. 1045, the composition range of which is 0.40 to 0.50 per cent carbon and 0.60 to 0.90 per cent manganese.

82. Property-mass Charts for 0.40 to 0.50 Per Cent Carbon Steel.—S.A.E. 1045 is the most frequently used of the medium-carbon steels. It is usually quenched in water and tempered at 540 to 650°C. (1000 to 1200°F.). The effect of mass on the mechanical properties of this steel has been determined by the

Society of Automotive Engineers[817] and the International Nickel Company.[700] The specimens used were taken longitudinally midway between the surface and the axis of bars of increasing diameter quenched from 815 to 845°C. (1500 to 1550°F.) for the smaller sizes or from 855 to 870°C. (1575 to 1600°F.) for the larger sizes.

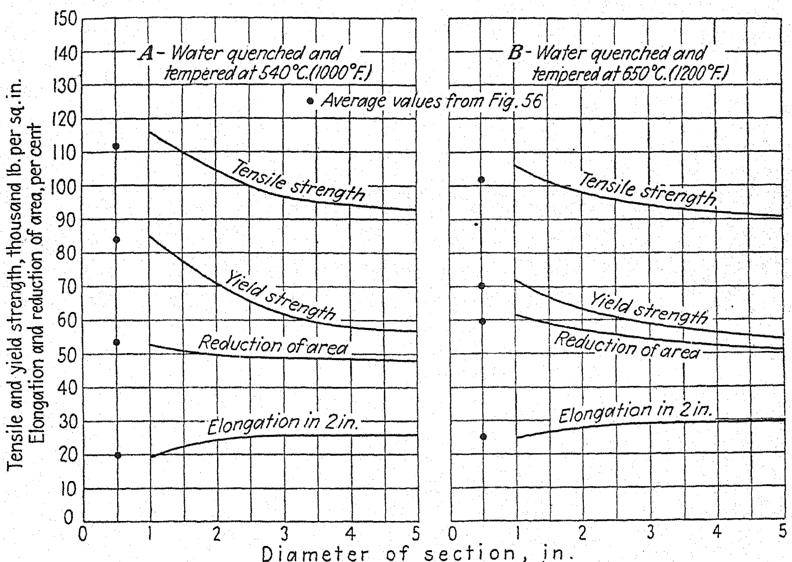

FIG. 65.—Effect of section size on the tensile properties of 0.40 to 0.50 per cent carbon, 0.60 to 0.90 per cent manganese (1045) steel. Specimens tested were 0.505 × 2 in., taken midway between surface and center of water-quenched and tempered bars of the sizes indicated. *(Society of Automotive Engineers,[817] and International Nickel Company.[700])*

The effect of mass on the properties of 1045 steel water quenched and tempered at 540 and 650°C. (1000 and 1200°F.), and on the same steel oil quenched and tempered at the same temperatures, is charted in Figs. 65 and 66. For the former the data of the Society of Automotive Engineers and the International Nickel Company, which checked closely, were averaged. In both charts the average values for steel of the same composition range, tempered at the same temperatures, taken from Fig. 56, page 200, are plotted by solid circles at the left of the curves.

Figures 65 and 66 show that the effect of mass is less evident when the material is oil quenched than when it is water quenched, and that for the same quenching treatment the effect of mass is

less evident when the sections are tempered at 650°C. (1200°F.) than when they are tempered at 540°C. (1000°F.).

The effect of section size on the properties of a steel containing 0.45 per cent carbon, 0.78 per cent manganese, 0.32 per cent silicon, 0.020 per cent sulphur, and 0.025 per cent phosphorus was determined on bars of $1\frac{1}{16}$ to 3-in. diameter by the Steel Research Committee(55) after quenching in water and oil from

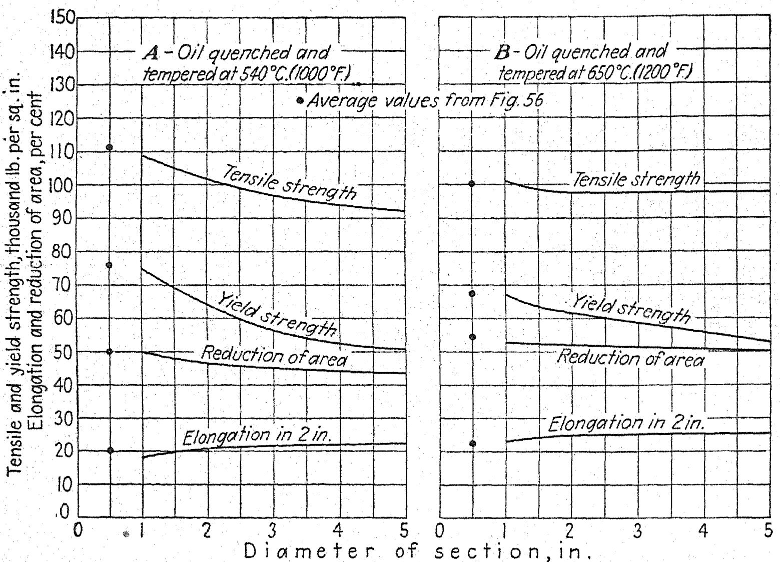

FIG. 66.—Effect of section size on the tensile properties of 0.40 to 0.50 per cent carbon, 0.60 to 0.90 per cent manganese (1045) steel. Specimens tested were 0.505 × 2 in., taken midway between surface and center of oil-quenched and tempered bars of the sizes indicated. (*Society of Automotive Engineers.*(817))

870°C. (1600°F.) and after water quenching followed by tempering at 500 and 600°C. (930 and 1110°F.). The results, plotted in Figs. 67 and 68, show the same trends as those in Figs. 65 and 66, except that the decrease in reduction of area with increase of section, pronounced in Figs. 65 and 66, is not confirmed by the data of the Steel Research Committee (Fig. 68).

The specimens treated by the Committee had markedly higher strength than the ones tested by the Society of Automotive Engineers and the International Nickel Company. This is evident from the following tabulation, which gives the strength properties of the water-quenched specimens tempered at 600°C.

(1110°F.), from Fig. 68*B*, and of those tempered at 540°C. (1000°F.), from Fig. 65*A*.

Property	1-in. section tempered at		2-in. section tempered at		3-in. section tempered at	
	600°C. (1110°F.)	540°C. (1000°F.)	600°C. (1110°F.)	540°C. (1000°F.)	600°C. (1110°F.)	540°C. (1000°F.)
Tensile strength, lb. per sq. in.	117,000	116,000	110,000	104,000	108,000	97,000
Elongation in 2 in., per cent	24	19	26	25	26	26
Reduction of area, per cent	61	53	61	50	59	49

Despite the fact that the Committee's specimens were tempered 60°C. (110°F.) higher than the specimens tested by the

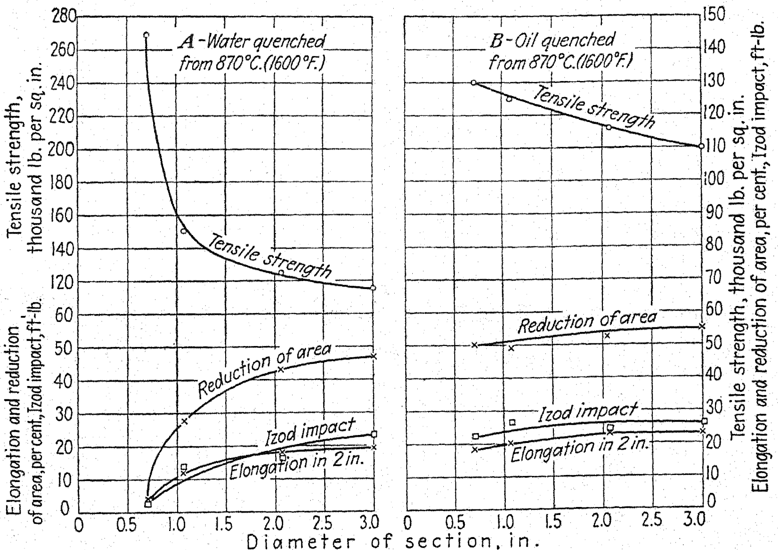

FIG. 67.—Effect of section size on the mechanical properties of a quenched and untempered 0.45 per cent carbon, 0.78 per cent manganese, 0.20 per cent silicon steel. Specimens tested were 0.564 × 2 in., cut from the core of round bars quenched in the sizes indicated. (*Steel Research Committee.*[55])

Society of Automotive Engineers and the International Nickel Company, the tensile strength of the Committee's specimens is appreciably higher, especially in the larger sections, and the

reduction of area is much higher in the specimens from all sections. This discrepancy in the data on the effect of mass should be cleared up by further work.

The marked decrease in Izod impact with section increase in the water-quenched and tempered specimens (Fig. 68*A* and *B*) and the much higher impact of the tempered small sections as compared with the untempered specimens (Fig. 68*A* and *B* and Fig. 67*A*) should be noted.

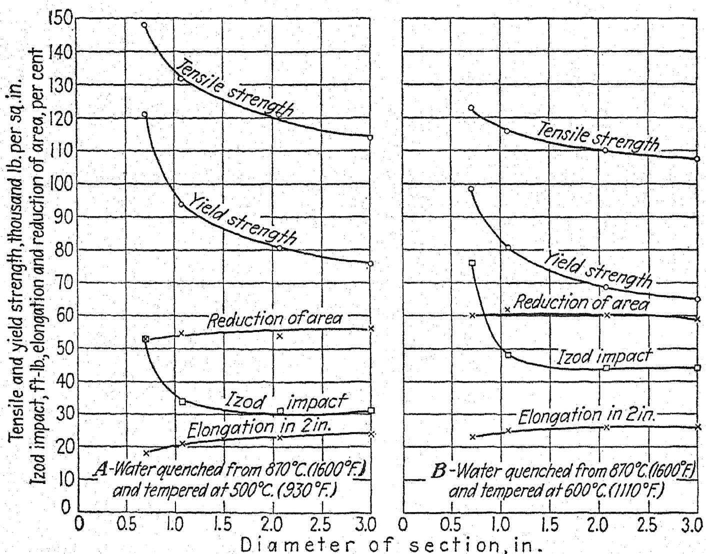

FIG. 68.—Effect of section size on the mechanical properties of a water quenched and tempered 0.45 per cent carbon, 0.78 per cent manganese, 0.20 per cent silicon steel. Specimens tested were 0.564 × 2 in., cut from the core of round bars heat treated in the sizes indicated. (*Steel Research Committee.*[55])

The effect of mass on the maximum strength and the proportional limit in torsion is shown in Fig. 69. The curves are for S.A.E. 1040 steel, water quenched and tempered at increasing temperatures.

83. Other Data on the Effect of Mass on Medium-carbon Steel.—Lake[731] investigated the tensile properties of longitudinal specimens cut from different locations in a 4.5-in. axle of a 0.40 per cent carbon steel. The location and size of the specimens are shown in Fig. 70 and the resulting properties in Table 53. All of the specimens were of standard dimensions (0.505 in. in diameter) except *A*, which had a diameter of 0.312 in. and was

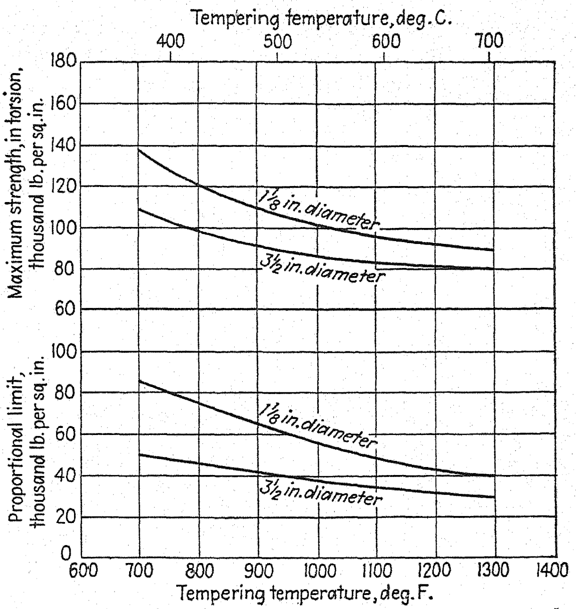

FIG. 69.—Effect of section size and tempering temperature on the torsional properties of 0.35 to 0.45 per cent carbon, 0.60 to 0.90 per cent manganese (1040) steel, water quenched from 830 to 870°C. (1525 to 1600°F.) in the sizes indicated. Specimen size not given. (*International Nickel Company.*[700])

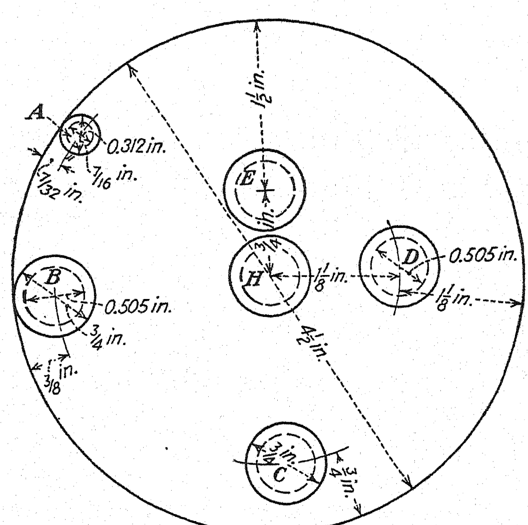

FIG. 70.—Sectional view of 4.5-in. axle housing showing location of test bars. (*Lake.*[731])

taken directly at the surface. Table 53 gives the properties of the axle as annealed, and as oil quenched after 1.5 hr. at 800°C. (1475°F.) and tempered 1 hr. at 400°C. (750°F.). In the annealed condition the size of section has little effect except possibly for the higher elongation values of the center specimens. In the oil-quenched and tempered specimens the effect of section is marked; the properties of the center specimen (*H*) are almost the same as the properties of the annealed specimen from the same location. Law[48] heated 18-in. cubes of a steel containing 0.34 per cent carbon, 0.19 per cent silicon, 0.74 per cent man-

TABLE 53.—EFFECT OF MASS ON TENSILE PROPERTIES OF A 4.5-IN. AXLE OF 0.40 PER CENT CARBON STEEL*

Specimen (see Fig. 70)	Annealed				Oil quenched and tempered at 400°C. (750°F.)			
	Tensile strength, lb. per sq. in.	Yield strength, lb. per sq. in.	Elongation in 2 in., per cent	Reduction of area, per cent	Tensile strength, lb. per sq. in.	Yield strength, lb. per sq. in.	Elongation in 2 in., per cent	Reduction of area, per cent
A	82,000	52,200	23	52	131,000	117,000	22	55
B	81,800	51,100	26	55	133,100	119,300	21	52
C	83,300	52,000	26	58	145,400	128,700	20	48
D	80,500	49,800	25	56	128,600	110,400	24	57
E	83,000	55,000	29	52	108,200	87,500	27	58
H	82,400	50,600	30	57	84,500	53,000	30	60

* Lake.[731]

ganese, 0.20 per cent nickel, 0.022 per cent sulphur, and 0.032 per cent phosphorus to 900°C. (1650°F.) and cooled them in air, in oil, in water, and by a water spray. Sections were cut from each cube, from face to opposite face, and tested for tensile and yield strengths, without tempering. Figure 71 shows the results.

In a recent study of deep-hardening abilities Kallen and Schrader[464] used two carbon steels and 12 low-alloy structural steels. As the carbon steels had almost identical compositions and hardening characteristics, the properties of only one of them are given here. This contained 0.45 per cent carbon, 0.79 per cent manganese, 0.3 per cent silicon, 0.022 per cent phosphorus, and 0.021 per cent sulphur. Round rods 20, 50, 100, and 200 mm. (0.79, 1.97, 3.94, and 7.87 in.) in diameter were used.

Test specimens were cut from the cores of all rods after heat treatment, and specimens were also cut from just below the surface of the rods 100 and 200 mm. in diameter. Samples were forged to the sections required; the difference in amount of hot working might have influenced the properties of the heat-treated material somewhat. All samples were normalized and

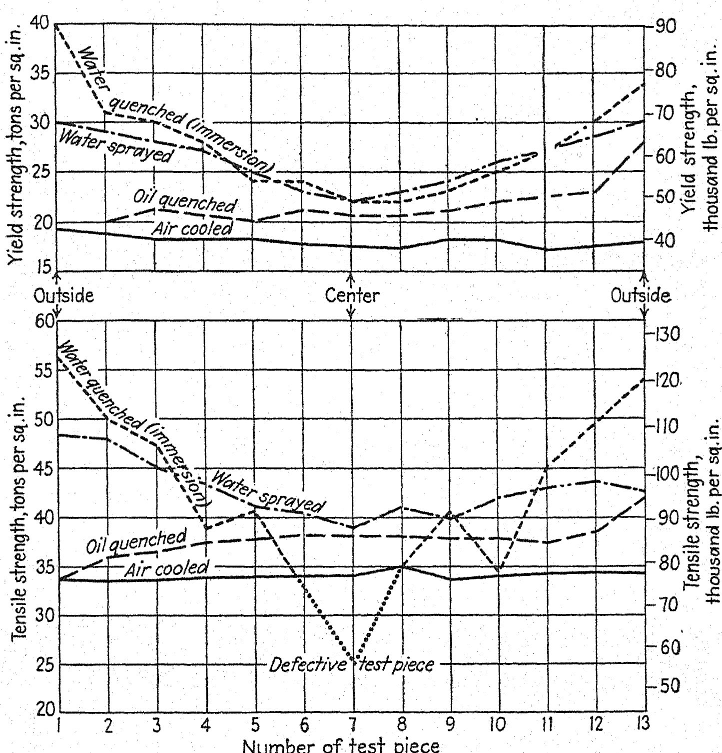

FIG. 71.—Tensile properties from surface to center of 18-in. cube containing 0.34 per cent carbon, 0.74 per cent manganese, 0.19 per cent silicon, and 0.20 per cent nickel, quenched from 900°C. (1650°F.) as shown and not tempered. (*Law.*[48])

then annealed before quenching. The Brinell-hardness values throughout sections of varying dimensions after quenching in water from 840°C. (1545°F.) and tempering at 550°C. (1020°F.) are shown in Fig. 72. Other properties of the heat-treated steel are shown in Table 54. The investigators concluded that the elastic ratio, ratio of yield strength to tensile strength, afforded the best measure of deep-hardening properties. As the data

in the table show, this ratio decreases as the section increases, which is due to the fact that the tensile strength remained substantially constant with increased section while the yield strength

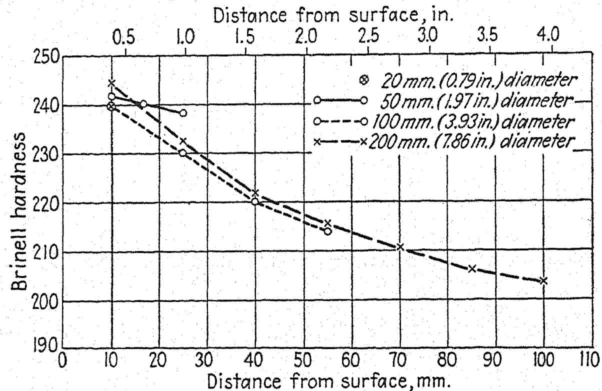

FIG. 72.—Effect of section size on the Brinell hardness of a steel containing 0.45 per cent carbon, 0.79 per cent manganese, and 0.3 per cent silicon, water quenched from 845°C. (1545°F.) and tempered at 550°C. (1020°F.). (*Kallen and Schrader.*[464])

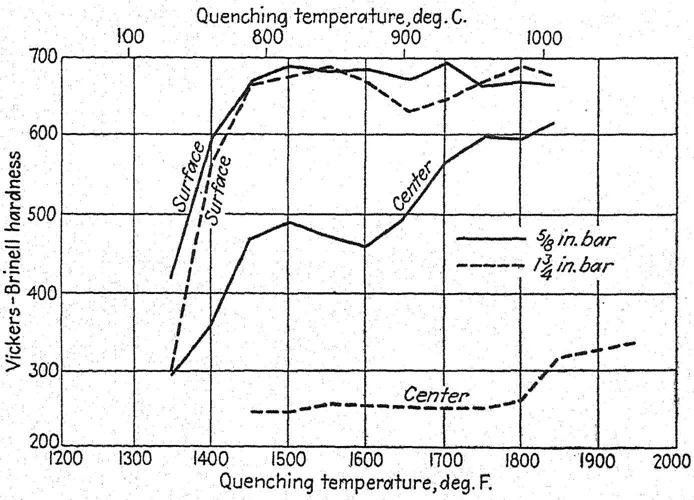

FIG. 73.—Effect of quenching temperature and mass on the Vickers-Brinell hardness of 0.30 to 0.40 per cent carbon (1035) steel. (*McMullan.*[271])

decreased appreciably. It is probably safe to conclude that the core of specimens over 20 mm. (0.79 in.) in diameter was not effectively hardened in quenching, and that the strength properties of the 20-mm. specimen were those of a quenched and

tempered steel, while those of the large specimens were of a normalized material.

The effect of quenching temperature and mass on the hardness of water-quenched S.A.E. 1035 steel was reported by McMullan.[271] Bars ⅝ and 1¾ in. in diameter were quenched and the Vickers-Brinell hardness was measured at surface and center. The values are shown in Fig. 73. Under the conditions of quenching—the specimens were agitated in the water—a high

TABLE 54.—PROPERTIES OF A STEEL CONTAINING 0.45 PER CENT CARBON, QUENCHED IN WATER FROM 840°C. (1545°F.) AND TEMPERED AT 550°C. (1020°F.)*

Diameter of treated specimen, mm.	Diameter of treated specimen, in.	Location of specimen	Tensile strength, lb. per sq. in.	Yield strength, lb. per sq. in.	Elastic ratio	Elongation ($l = 5\,d$), per cent	Reduction of area, per cent	Charpy impact value, m-kg. per sq. cm.
20	0.79	Core	113,000	97,000	0.86	19.5	62	15
50	1.97	Core	110,000	75,500	0.69	21.5	58	7.5
100	3.94	Surface	110,000	71,000	0.65	20	56	8
100	3.94	Core	105,000	68,000	0.65	23	56	7.5
200	7.87	Surface	114,000	72,000	0.63	16	54	5
200	7.87	Core	105,000	64,000	0.61	23.5	56	4.5

* Kallen and Schrader.[464]

quenching temperature produced a relatively great increase in hardness at the center of the ⅝-in. bar but had little effect upon the 1¾-in. bar.

84. Heavy Sections of Medium-carbon Steel.—In discussion of a paper by Greene[255] Fry summarized the properties of large locomotive forgings made from 26 heats of acid open-hearth steel containing 0.42 to 0.52 per cent carbon, 0.53 to 0.69 per cent manganese, 0.19 to 0.32 per cent silicon, 0.024 to 0.037 per cent sulphur, and 0.027 to 0.045 per cent phosphorus. The ingots were pressed to billets, the forgings made, normalized, and annealed. The crankpins and piston rods from which the specimens were taken ranged from 5⅜ in. to 8¾ in. in diameter and the connecting rods were 3 or 5 in. by 10 or 11 in. The driving axles were 10¾ to 11½ in. in diameter. Fry's values are given in Table 55; normalizing and annealing temperatures were not given.

MacPherran and Harper,[129] in a report on the effect of a spheroidizing anneal on the properties of large forgings, gave the data reproduced in Table 56. The heat treatments given to the forgings from which the specimens were cut were as follows:

Number	Treatment
1*A*........	Annealed 5 hr. at 795°C. (1465°F.) and furnace cooled
1*B*........	Same as 1*A*, followed by spheroidizing 36 hr. at 675°C. (1250°F.) and furnace cooling
2*A*........	Annealed 8 hr. at 800°C. (1475°F.) and furnace cooled
2*B*........	Same as 2*A*, followed by spheroidizing 56 hr. at 675°C. (1250°F.) and furnace cooling
3..........	Annealed 8 hr. at 815°C. (1500°F.) and furnace cooled; reheated 4 hr. at 700°C. (1290°F.) and furnace cooled

The spheroidizing treatment lowered the tensile and yield strengths and increased greatly the values for elongation and reduction of area.

The study by Maurer and Korschan on the mechanical properties of large sections of low-carbon steel (see page 217) was continued by Maurer and Gummert[738] who used a medium-carbon steel, melted in two basic furnaces, containing about 0.30 per cent carbon, 0.25 per cent silicon, and 0.65 per cent manganese

TABLE 55.—MECHANICAL PROPERTIES OF NORMALIZED AND ANNEALED ACID OPEN-HEARTH STEEL CONTAINING 0.42 TO 0.52 PER CENT CARBON AND 0.53 TO 0.69 PER CENT MANGANESE*

Property	Crankpins, connecting rods, piston rods. Less than 9 in. in diameter or thickness; 43 tests			Driving axles. Over 9 in. in diameter; 17 tests		
	Average	Maximum	Minimum	Average	Maximum	Minimum
Yield strength, lb. per sq. in...	54,000	56,000	51,000	54,300	56,000	50,500
Tensile strength, lb. per sq. in...	90,300	96,000	85,000	90,700	93,500	85,500
Elongation in 2 in., per cent..	25.5	28.0	23.5	24.8	26.5	22.0
Reduction of area, per cent....	43.3	52.0	37.5	40.3	45.0	37.5

* Fry, in discussion of paper by Greene.[255]

TABLE 56.—MECHANICAL PROPERTIES OF LARGE FORGINGS AS ANNEALED AND AS SPHEROIDIZED*

Number	Composition, per cent					Size of ingot, in.	Diameter of forging, in.	Direction of specimen	Properties			
	C	Mn	Si	S	P				Tensile strength, lb. per sq. in.	Yield strength, lb. per sq. in.	Elongation,† per cent	Reduction of area, per cent
1*A*	0.47	0.75	0.24	0.041	0.038	18	8½	Longitudinal	104,500	51,500	18	28
1*B*									88,000	43,000	26	46
2*A*	0.45	0.53	0.17	0.028	0.027	31	24⅛	Radial	84,000	38,500	18	26
2*B*									66,000	32,000	27	42
3	0.46	0.58	0.16	0.027	0.014	42	30	‡	77,500	35,500	27	44

* MacPherran and Harper.[129]
† Gage length not given.
‡ Longitudinal, tangential, and radial values all agreed closely.

with sulphur at about 0.02 per cent and phosphorus slightly lower. Approximately 0.12 per cent nickel and 0.02 per cent chromium were also present. A carbon-manganese, a nickel, and a nickel-chromium-molybdenum steel were also studied, as in the previous work.

The 45-ton ingots [1430 mm. (56 in.) base, 1530 mm. (60 in.) top, 2800 mm. (110 in.) high] were forged down to sections 650, 840, and 1020 mm. (25.6, 33, and 40.2 in.) in diameter, equivalent to an elongation of 2, 3, and 5 times. The tests were carried on as in the previous work (see page 217). The results are shown in Figs. 74 and 75.

In general, in the large forgings of 0.30 per cent carbon steel, as in the large forgings of 0.17 per cent carbon steel discussed previously (page 221), there is a tendency for the strength to decrease from surface to core. The properties of the longitudinal specimens are generally better than the properties of the transverse specimens similarly treated; but the transverse properties are considerably better and in general more uniform than is evident in some of the other work on large forgings. Doubtless the cleanness of the steel is an important factor in the transverse properties of large sections.

Maurer and Gummert called attention to the fact, which is evident from a study of Figs. 74 and 75, that there is little difference in properties of the forging which was reduced 80 per cent and the forging which was reduced 50 per cent. These investigators also attempted to trace the relation between the grain size and properties of their 45-ton ingot and the forgings made from it and the grain size and properties of the 100-ton ingot used by Maurer and Korschan. They found that there is a considerable difference in grain size between rim and core in both ingots and in the forgings made from these ingots but that this difference is almost eliminated by air quenching and is further decreased by oil quenching. Regardless of the heat treatment, however, there is a difference in grain size between the 45-ton and the 100-ton ingots which persists through the forging operation. Because of the large difference in carbon (0.30 per cent versus 0.17 per cent), it is difficult to detect from Maurer and Gummert's curves (not reproduced here) if this difference in actual grain size has any direct relation to the properties.

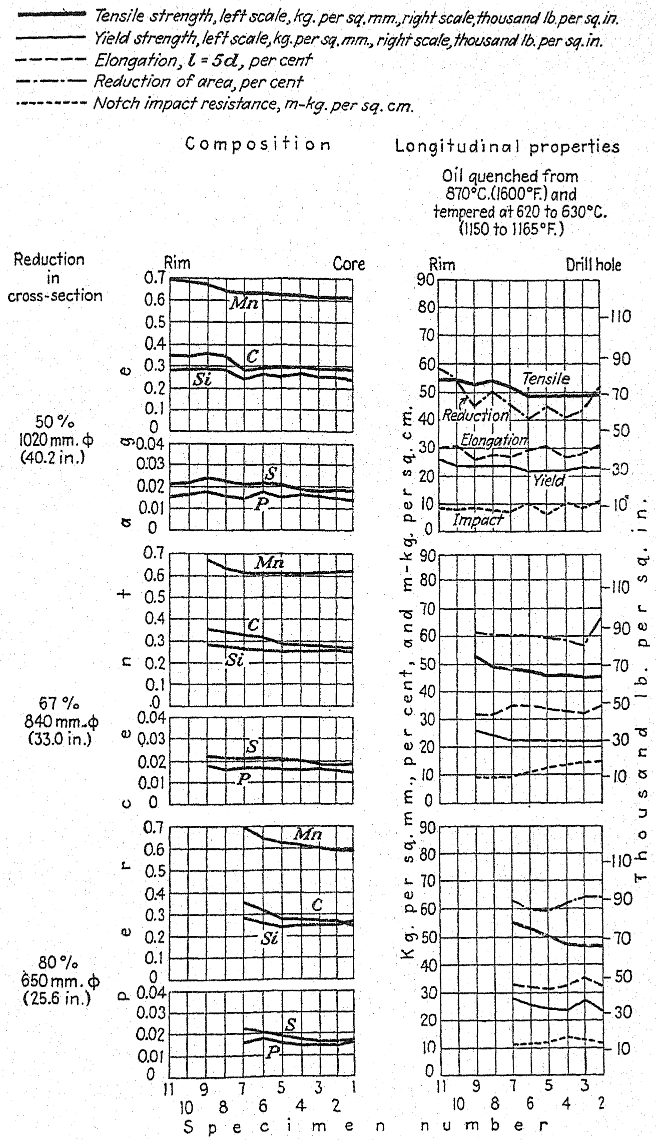

FIG. 74.—Composition and longitudinal properties throughout the cross-section of large forgings with an average composition of 0.30 per cent carbon, 0.65 per cent manganese, and 0.25 per cent silicon. Each specimen number represents a section of about 1.5 in., beginning at core of forging. (*Maurer and Gummert.*[738])

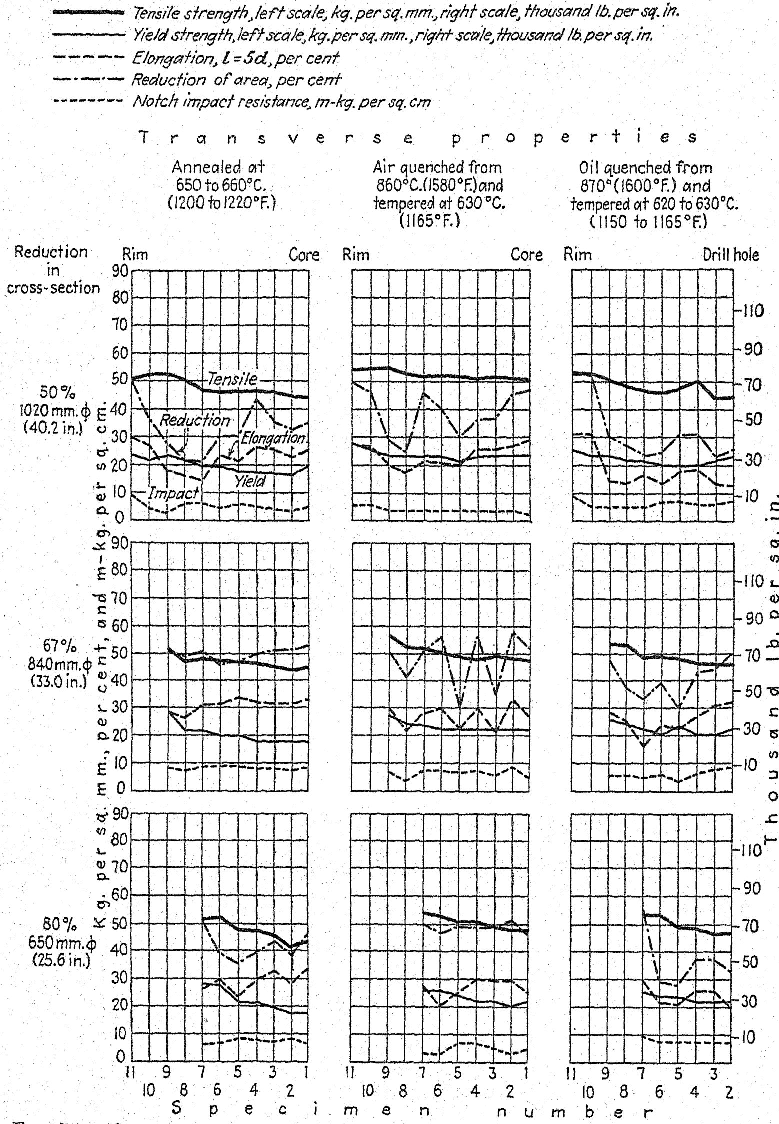

FIG. 75.—Transverse properties of large forgings whose composition and longitudinal properties are given in Fig. 74. (*Maurer and Gummert.*[738])

C. EFFECT OF CROSS-SECTION ON MECHANICAL PROPERTIES OF HIGH-CARBON STEELS

Although practically all high-carbon steels are used in the heat-treated condition there are few data on the tensile properties of small sections and even fewer on the effect of mass. This is to be expected; as was emphasized in the last chapter, the tensile properties of commercial iron-carbon alloys containing more than about 0.65 per cent carbon are of minor practical importance as compared with those properties by which one may judge the fitness of a high-carbon steel for tools, dies, and the like. The few data on the tensile properties of these steels as quenched and tempered are summarized on pages 204 to 211. The fewer data on the effect of mass are given below.

85. Factors Complicating the Mass Effect in the Heat Treatment of High-carbon Steels.—In the first volume of this monograph Epstein discussed at length the variables which affect the transformation of austenite to martensite and the structural and other changes which take place when a quenched (martensitic) carbon steel is tempered. Included in this discussion were data on the effect of mass (Volume I, pages 286 to 294). It is well known that, with the usual conditions of heat treatment, even relatively small sections of high-carbon steel do not harden throughout. Even with the most drastic quenching, the hardened layer on a 1-in. round bar may be no more than ⅛ or ¼ in. thick. Not only is there this well-known mass effect, but there is also, as Epstein showed (page 188), a variation in the hardenability of carbon steels of apparently the same composition and cross-section, treated in the same way. Epstein reviewed the work of Bain and others in this field and concluded that, tentatively at least, differences in hardenability may be attributed chiefly to oxides and grain size. The latter seems to be the most potent single factor in increasing the hardening capacity, and this in turn seems to be controlled by finely dispersed particles, presumably stable oxides such as aluminum oxide.

For certain applications, shallow hardening may actually be advantageous because it results in the production of parts having a hard wear-resistant surface and a softer but tougher core. For other applications, however, as in springs and some small tools, it is essential that the steel harden throughout; thus, for such

parts attention may have to be given to the inherent hardenability as well as to the effect of mass.

86. Data on the Effect of Mass on High-carbon Steels.—Because of the great difference in the inherent hardenability of high-carbon steels of the same composition and cross-section it is obviously impossible to discuss adequately the influence of section on the properties of these steels, even if any data were available. This is shown clearly by Fig. 76 from Bain,[508]

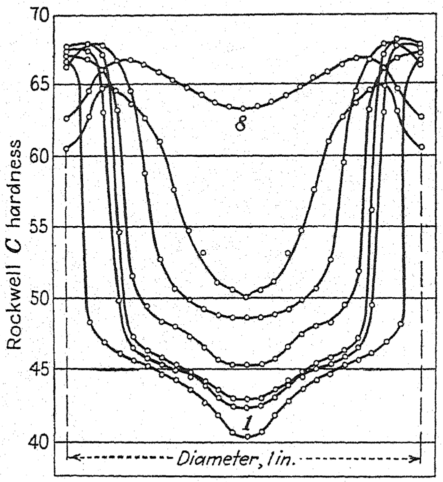

Fig. 76.—Hardness of the cross-section of eight 1-in. diameter round bars of 1 per cent carbon tool steel given the same hardening treatment. (*Bain.*[508])

who explored the hardness across 1-in. diameter bars of eight 1 per cent carbon steels of almost identical composition (as determined by ordinary chemical analysis). The sections were treated in the same way. Seven of the steels were shallow hardening and only one, steel No. 8, hardened throughout the entire section.

J. A. Mathews,* in commenting upon a first draft of this chapter, pointed out that in massive sections, as for example in high-carbon die blocks, there is a certain degree of compensation for the effect of mass in the quenching and tempering. While the effect of quenching is to produce different degrees of hardness, highest at the corners, lower at the centers of the faces, and lowest at the center of the whole block, in tempering, the corners will soften faster than the centers of the faces so that eventually,

* Private communication.

at around 260 to 315°C. (500 to 600°F.), practically equal hardness will be obtained over the faces. Less drastic hardening, as quenching in oil, does not give the same results as quenching in water followed by tempering to the same hardness that would be obtained by the oil quenching, owing to retained austenite. This can be shown by studying the changes in induction and in coercive force.

D. EFFECT OF GRAIN SIZE AND OTHER VARIABLES

Since the pioneer research by McQuaid and Ehn reported in 1922, many workers (almost exclusively in the United States) have investigated the effect of grain size, especially in its relation to carburizing, hardenability, machinability, and impact properties. This work has been reviewed and summarized by Epstein in Volume I of this monograph (pages 181 to 191, 371 to 383). The general conclusions which may be drawn from this large volume of work are: (1) coarse-grained steels harden more deeply, carburize more deeply, and machine more readily than fine-grained steels; and (2) fine-grained steels have a wider heating range in heat treatment, are less subject to cracking in quenching, and are tougher, especially at low temperatures, than coarse-grained steels.

Less attention has been paid to the relation between grain size and mechanical properties; it was not until the grain-size Symposium was held by the American Society for Metals in 1934 that enough data appeared to enable this relationship to be traced.

As is well known, steels differ according to their actual grain size, and according to their tendency toward grain growth (inherent grain size).* By the former is meant the grain size

* The tendency of a steel toward grain growth—the grain-growth behavior —over a range of temperatures has been almost universally termed "inherent grain size" in the literature. Although this term is somewhat misleading, it is convenient; hence it will probably be difficult to supplant it by another expression. In this connection, E. S. Davenport wrote (private communication): "While a loose terminology has grown up around this subject whereby steels are referred to merely as being either coarse grained or fine grained (and everyone understands what is meant), the terms are not strictly correct. Steels are really coarse grained or fine grained only insofar as they have been above or below their particular coarsening temperatures. It would be more correct to speak of steels with high or low coarsening

which depends upon the temperature to which the steel is heated for heat treatment and upon the time the steel is held at that temperature. If other things are equal, the higher the temperature and the longer the time, the larger will be the resulting grains (actual grain size). By inherent grain-growth tendency is meant the tendency of the austenite grains to grow at temperatures above the critical. Again if other things are equal, the austenite grains of inherently coarse-grained steel start to grow at temperatures just above the critical; the austenite grains of an inherently fine-grained steel, on the contrary, do not grow appreciably until the material is heated considerably above the critical range, for example to 980°C. (1800°F.).

87. Data on the Relation between Inherent Grain-growth Tendency and Mechanical Properties.—In an investigation made primarily to determine methods of controlling the inherent grain-growth tendency by adding aluminum in the melting process, Epstein, Nead, and Washburn[695] determined the tensile and impact properties of coarse- and fine-grained basic open-hearth steels. The ingots were rolled to heavy billets, and one billet from the middle and the bottom of an ingot of each heat was forged to a 1-in. round. To bring out the inherent differences in the tendency toward grain growth these 1-in. rounds were heated to 885 and 955°C. (1625 and 1750°F.), held 1 hr., and air cooled. Tensile specimens, 0.505 × 2 in., and Charpy impact specimens (keyhole notch) were machined from the center of the air-cooled bars. Epstein, Nead, and Washburn did not give their complete data; the average values for the low- and medium-carbon heats are reproduced in Table 57. From these data they concluded that:

. . . the yield strength of the finer-grained steel was appreciably higher than that of the coarser-grained steel, this being especially

temperatures. Viewed in this light, the term inherent grain size has little or no meaning; in fact, it is misleading since it implies that a steel has some *particular* grain size which is characteristic of it. If the word inherent must be used, it seems as though it should be coupled with the word tendency or behavior or some similar phrase to indicate the dependence of austenitic grain size upon temperature." In accord with Davenport's suggestion, the tendency of a steel toward grain growth has, so far as possible, been termed "inherent grain-growth tendency" in this chapter. No confusion should result therefrom for those readers accustomed to think of this tendency as "inherent grain size."

TABLE 57.—AVERAGE MECHANICAL PROPERTIES OF COARSE- AND FINE-GRAINED CARBON STEELS*

Carbon range,† per cent	Air cooled from		Location of specimen in ingot	Coarse-grained (2 to 3)‡					Fine-grained (5 to 7)‡				
	°C.	°F.		Tensile strength, lb. per sq. in.	Yield strength, lb. per sq. in.	Elongation in 2 in., per cent	Reduction of area, per cent	Charpy impact, ft-lb.	Tensile strength, lb. per sq. in.	Yield strength, lb. per sq. in.	Elongation in 2 in., per cent	Reduction of area, per cent	Charpy impact, ft-lb.
0.15 to 0.19	885	1625	Middle	61,200	39,400	35.4	64.6	43.4	61,000	42,000	36.0	66.1	47.6
	885	1625	Bottom	60,300	37,300	35.9	66.1	45.0	60,500	41,500	36.4	67.3	47.5
0.15 to 0.19	955	1750	Middle	60,600	32,600	32.0	62.2	40.5	60,300	39,100	35.5	66.9	45.5
	955	1750	Bottom	59,500	33,200	34.0	62.4	41.0	59,500	38,600	33.8	64.9	46.2
0.32 to 0.43	885	1625	Middle	86,700	49,100	28.3	56.9	24.7	82,000	51,100	31.0	58.1	28.0
	885	1625	Bottom	82,800	44,200	30.0	53.4	26.2	78,700	46,800	30.0	58.3	31.0
0.32 to 0.43	955	1750	Middle	90,000	45,500	25.3	49.4	21.8	85,300	47,200	29.0	54.0	27.6
	955	1750	Bottom	86,100	43,700	26.3	51.7	27.6	81,300	47,800	29.3	56.1	31.3

* Epstein, Nead, and Washburn.(695)

† Low-carbon steels, average of five heats; medium-carbon steels, average of three heats.

‡ A.S.T.M. grain-size chart.(587)

noticeable after the 955°C. (1750°F.) treatment which brought out a decided difference in grain size between the two types of steel. After this treatment, in the steels ranging from 0.15 to 0.19 per cent carbon the yield strength was about 5000 lb. per sq. in. higher in the fine- than in the coarse-grained steel. After the 885°C. (1625°F.) treatment, however, when the difference in grain size between the coarse- and fine-grained steels was less, there was a less difference in yield strength. In the as-rolled condition there was practically no difference. The tensile strength, however, did not differ between the coarse- and fine-grain steels, except in the higher carbon range, where the coarse-grain steel showed a somewhat higher tensile strength. The hardness in general conformed to the tensile strength. The elongation and reduction of area were slightly higher in the fine-grain steels. Although impact tests did not show the large differences between the coarse- and fine-grain steels expected, they were definitely higher in the finer-grain steels, particularly after the 955°C. (1750°F.) treatment.

Schane,(764) in the same symposium, reported a series of tests made on coarse-grained and on fine-grained 1040 steel containing 0.40 per cent carbon, 0.75 per cent manganese, 0.20 per cent silicon, and less than 0.035 per cent sulphur and phosphorus. The material was treated in the form of 1-in. hot-rolled rounds from the center of which 0.505 × 2-in. specimens were machined. The treatments and the resulting properties are given in Table 58.

No conclusions were drawn from these results by Schane except the statement that:

. . . in the ordinary heat-treating ranges the hardenability and grain structure of this [fine-grained] type are uniform whereas to obtain this same condition in the coarser grained types it is necessary to go to much higher temperatures which are usually impractical (on account of scaling, decarburization, etc.) and produce less desirable properties.

It is evident from Table 58 that, for the quenching treatments used, the tensile strength is generally lower and the elongation and reduction of area are somewhat higher in the fine-grained material. The impact resistance is much higher. In the normalized and in the annealed specimens the yield strength of the coarse- and fine-grained steels is about the same, in the quenched and tempered specimens the fine-grained steel has a somewhat lower yield.

The properties of coarse-grained and of fine-grained 1040 steel showing the effect of quenching and tempering temperatures

TABLE 58.—MECHANICAL PROPERTIES OF COARSE- AND FINE-GRAINED 0.40 PER CENT CARBON, 0.75 PER CENT MANGANESE STEEL*

Treatment				Coarse grained					Fine grained				
Temperature		Held, hr.	Cooled in	Tensile strength, lb. per sq. in.	Yield strength, lb. per sq. in.	Elongation in 2 in., per cent	Reduction of area, per cent	Izod impact, ft-lb.	Tensile strength, lb. per sq. in.	Yield strength, lb. per sq. in.	Elongation in 2 in., per cent	Reduction of area, per cent	Izod impact, ft-lb.
°C.	°F.												
760	1400	1	Air	96,000	53,000	26.0	52.2	28	87,800	55,000	30.0	59.3	69
815	1500	1	Air	97,000	56,000	23.5	48.0	27	88,800	54,000	29.0	58.0	56
870	1600	1	Air	99,000	52,000	22.0	45.0	15	89,000	52,000	28.0	57.0	55
925	1700	1	Air	101,700	53,000	21.5	43.6	13	88,800	54,000	27.5	56.2	55
980	1800	1	Air	103,400	52,000	20.0	40.6	9	89,000	54,000	27.0	52.7	48
1040	1900	1	Air	102,700	52,000	20.0	39.4	9	94,500	52,000	24.0	48.5	31
1095	2000	1	Air	102,100	53,000	20.0	39.7	6	94,000	50,000	25.0	49.4	30
705	1300	14	Furnace	77,100	44,400	34.0	58.1	20	74,900	45,400	32.0	61.4	43.34
760	1400	..	Water†	98,200	59,670	22.0‡	58.0	7.9	101,900	66,900	26.0‡	63.8	65.89
790	1450	..	Water†	115,600	80,300	20.5‡	56.7	12.20	107,300	78,900	26.0‡	65.4	90.95
815	1500	..	Water†	116,900	81,300	20.0‡	55.9	23.26	109,800	76,900	23.0‡	64.5	90.92
845	1550	..	Water†	117,600	82,100	19.0‡	51.1	7.7	108,900	80,400	23.0‡	64.8	79.85
870	1600	..	Water†	119,600	85,300	19.0‡	53.3	11.17	108,900	77,400	23.0‡	64.0	76.79

* Schane.(764)

† Followed by tempering at 565°C. (1050°F.).

‡ In 8 in.

TABLE 59.—MECHANICAL PROPERTIES OF COARSE- AND FINE-GRAINED 0.40 PER CENT CARBON, 0.73 PER CENT MANGANESE STEEL*

Quenching temperature		Cooled in	Tempering temperature		Coarse grained					Fine grained				
°C.	°F.		°C.	°F.	Tensile strength, lb. per sq. in.	Yield strength, lb. per sq. in.	Elongation in 2 in., per cent	Reduction of area, per cent	Izod impact,† ft-lb.	Tensile strength, lb. per sq. in.	Yield strength, lb. per sq. in.	Elongation in 2 in., per cent	Reduction of area, per cent	Izod impact,† ft-lb.
870	1600	Air	. . .	. . . .	98,200	53,600	20.5	44.8	10	91,300	59,900	24.0	50.5	80
760	1400	Water	565	1050	100,300	61,500	22.0	58.0	10	100,400	65,900	24.5	54.2	80
790	1450	Water	565	1050	117,300	81,400	21.0	57.0	17	107,000	76,100	24.5	54.7	88
815	1500	Water	565	1050	117,100	82,000	20.5	56.0	27	110,000	77,000	22.0	54.0	87
845	1550	Water	565	1050	118,300	83,000	19.5	52.2	8	108,000	78,900	22.5	55.0	80
870	1600	Water	565	1050	120,000	86,600	19.0	52.8	15	106,000	77,000	22.0	56.2	78
830	1525	Water	425	800	131,500	102,600	16.0	42.5	10	128,300	98,300	20.0	51.0	68
830	1525	Water	480	900	129,600	97,800	17.5	46.0	11	122,900	93,000	20.5	51.0	69
830	1525	Water	540	1000	120,000	85,200	20.5	51.0	15	115,100	80,100	21.0	53.5	70
830	1525	Water	595	1100	107,200	80,400	21.0	55.0	26	102,400	72,300	22.5	58.5	80
830	1525	Water	650	1200	100,800	75,800	21.0	57.5	30	98,700	72,100	25.0	60.5	90

* Rolf.(759)

† Average of five tests.

were reported by Rolf.[759] The heats contained 0.40 per cent carbon, 0.72 and 0.74 per cent manganese, 0.19 and 0.21 per cent silicon, less than 0.040 per cent sulphur, and less than 0.025 per cent phosphorus. One-inch round bars were heated to the temperatures shown in Table 59 and held 1 hr., being cooled as shown. The tensile specimens were of standard dimensions (0.505 × 2 in.), and the impact values were obtained on standard V-notch specimens. The properties are given in Table 59.

For all the treatments the tensile and yield strengths were slightly lower and the elongation and reduction of area values higher in the fine-grained specimens. Impact values of these specimens were very much higher.

88. Comparison of Mechanical Properties of Coarse- and Fine-grained Steels.—Because of the recent widespread interest in grain-size control in carbon and alloy steels, it will be of advantage to summarize the data given in the investigations described in the previous section, so that, if any marked changes in properties with grain size are shown by these reports, they will be more readily apparent.

This was done with the data in Tables 57, 58, and 59; the results are shown in Table 60. In this compilation the mechanical properties of the fine-grained steels were compared with the properties of the coarse-grained steels for each investigator and for most of the treatments (no values are given for specimens treated at 760°C. (1400°F.) or below) on a percentage basis. This has the advantage that it eliminates any effect of different methods of determining yield strength or impact resistance, and any differences in actual elongation due to a difference in gage length. The values shown in Table 60 should, consequently, be directly comparable unless the heat treatment used by an investigator for a coarse-grained specimen varied inadvertently from that used for the corresponding fine-grained specimen, or unless there were unobserved flaws in a specimen which affected individual results.

As it is not likely that these sources of error were present to an important extent, the percentages in Table 60 should show with fair accuracy the effect of grain size.* Parenthetically it should be stated that variations in a property of less than 3

* Neither Schane nor Rolf gave grain-size numbers for the specimens they used. An inspection of the micrographs of their steels indicated,

TABLE 60.—PERCENTAGE INCREASE OR DECREASE IN MECHANICAL PROPERTIES OF FINE-GRAINED STEELS AS COMPARED WITH COARSE-GRAINED STEELS OF APPROXIMATELY THE SAME COMPOSITION AND HEAT TREATMENT

Investigator	Approximate carbon content, per cent	Treatment	Tensile strength	Yield strength	Elongation	Reduction of area	Impact value
Epstein, Nead, and Washburn[695]	0.17	Air cooled from 885°C. (1625°F.)	0	+ 9	0	0	+ 8
		Air cooled from 955°C. (1750°F.)	0	+15	+ 5	+ 6	+ 13
Epstein, Nead, and Washburn[695]	0.37	Air cooled from 885°C. (1625°F.)	− 5	+ 5	+ 5	+ 6	+ 16
		Air cooled from 955°C. (1750°F.)	− 5	+ 7	+13	+ 9	+ 19
Schane[764]	0.40	Air cooled from 815 and 870°C. (1500 and 1600°F.)	− 9	0	+25	+24	+164
		Air cooled from 925 and 980°C. (1700 and 1800°F.)	−13	0	+31	+30	+368
		Air cooled from 1040 and 1095°C. (1900 and 2000°F.)	− 8	0	+23	+24	+307
Schane[764]	0.40	Water quenched from 790 and 815°C. (1450 and 1500°F.)*	− 7	− 3	+21	+15	+413
		Water quenched from 845 and 870°C. (1550 and 1600°F.)*	− 8	− 6	+21	+23	+737
Rolf[759]	0.40	Water quenched from 790 and 815°C. (1450 and 1500°F.)*	− 8	− 6	+12	− 4	+300
		Water quenched from 845 and 870°C. (1550 and 1600°F.)*	−10	− 8	+15	+ 6	+590
Rolf[759]	0.40	Water quenched from 830°C. (1525°F.) and tempered at 425°C. (800°F.)	0	− 4	+25	+20	+580
		Water quenched from 830°C. (1525°F.) and tempered at 480°C. (900°F.)	− 4	− 5	+17	+11	+527
		Water quenched from 830°C. (1525°F.) and tempered at 540°C. (1000°F.)	− 4	− 6	0	+ 5	+366
		Water quenched from 830°C. (1525°F.) and tempered at 595°C. (1100°F.)	− 4	−10	+ 7	+ 6	+208
		Water quenched from 830°C. (1525°F.), and tempered at 650°C. (1200°F.)	0	− 5	+19	+ 5	+200

* Followed by tempering at 565°C. (1050°F.).

per cent were considered to be within the experimental error; these have, therefore, been called zero in Table 60.

In commenting on the data of Table 60, it should be stated first that the variation in trend between the values by Schane, by Rolf, and by Epstein, Nead, and Washburn is sufficiently marked so that additional data should be obtained before the conclusions given below can be considered definite.

With this proviso clearly in mind, it may be stated tentatively, for steels containing about 0.40 per cent carbon and 0.75 per cent manganese and for a definite heat treatment, that:

1. Fine-grained steels have a slightly lower tensile strength and the same or slightly lower yield strength than coarse-grained steels.
2. Fine-grained steels usually have markedly higher (10 to 25 per cent) elongation and reduction of area than coarse-grained steels.
3. The impact resistance at normal temperature for a steel with grain size 5 to 7 is from two to six times the impact resistance of a corresponding steel with a grain size of 2 or 3.

89. Effect of Aging at Normal Temperatures on the Properties of Quenched High-carbon Steel.—The changes in structure and physical properties which take place when commercial iron-carbon alloys are aged at or slightly above room temperature have been discussed in some detail in Volume I of this monograph (pages 384 to 396). In that discussion Epstein divided aging phenomena into three classes, (1) quench aging, (2) strain aging, and (3) blue-heat phenomena, and gave a few data on mechanical properties, especially on hardness, to illustrate the changes which take place. All the data quoted by Epstein to illustrate the effects of quench aging were on low-carbon steels in which this form of aging is most pronounced and best known. Equally well known are the effects of strain aging on the mechanical properties, touched upon briefly by Epstein and discussed in more detail on page 152 in the chapter on cold-worked steels.

It is not so generally recognized, however, that changes in properties also occur in quenched high-carbon steels if these are held for relatively long periods of time at or just above room temperature. Scott[199] found that minute dimensional changes may take place with time in quenched and quenched and tem-

however, that the coarse-grained material corresponded to Nos. 2 and 3, and the fine-grained steel corresponded to Nos. 5 to 7 of the A.S.T.M. chart.

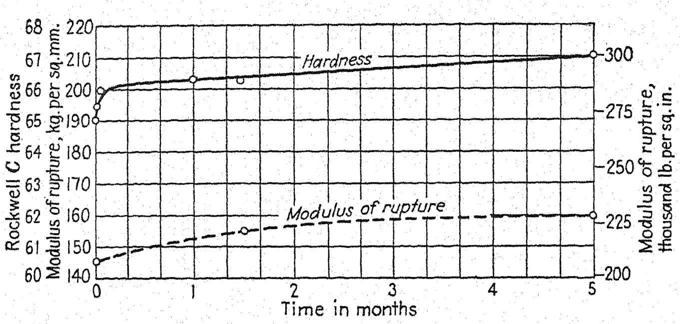

Fig. 77.—Effect of aging for 5 months at room temperature on the Rockwell *C* hardness and modulus of rupture of water-quenched high-carbon steel containing 0.98 per cent carbon and 0.43 per cent manganese. (*Steinberg and Subow.*[493])

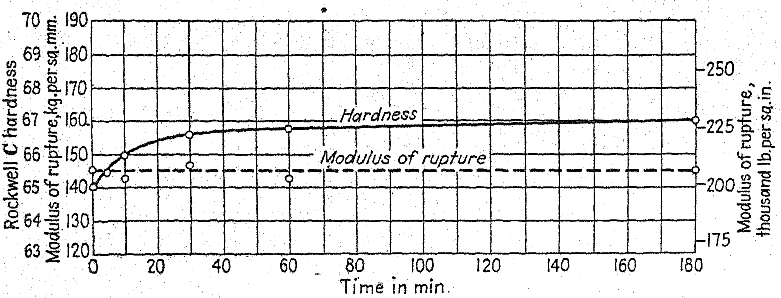

Fig. 78.—Effect of aging for 180 min. at 75°C. (165°F.) on the Rockwell *C* hardness and modulus of rupture of water-quenched high-carbon steel containing 0.98 per cent carbon and 0.43 per cent manganese. (*Steinberg and Subow.*[493])

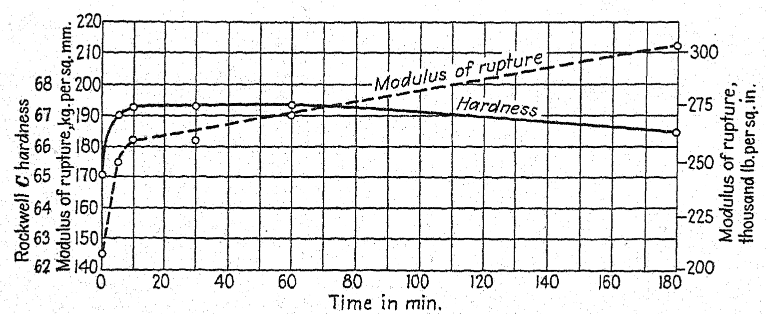

Fig. 79.—Effect of aging for 180 min. at 100°C. (210°F.) on the Rockwell *C* hardness and modulus of rupture of water-quenched high-carbon steel containing 0.98 per cent carbon and 0.43 per cent manganese. (*Steinberg and Subow.*[493])

pered high-carbon steels as used for tools and gages. This is a subject which is too specialized to be discussed here. Of more general importance is the change in mechanical properties with time; this has received attention.

Steinberg and Subow[493] quenched a steel containing 0.98 per cent carbon, 0.43 per cent manganese, and 0.19 per cent silicon in water from 750°C. (1380°F.) and determined the hardness and modulus of rupture* immediately and after holding for various

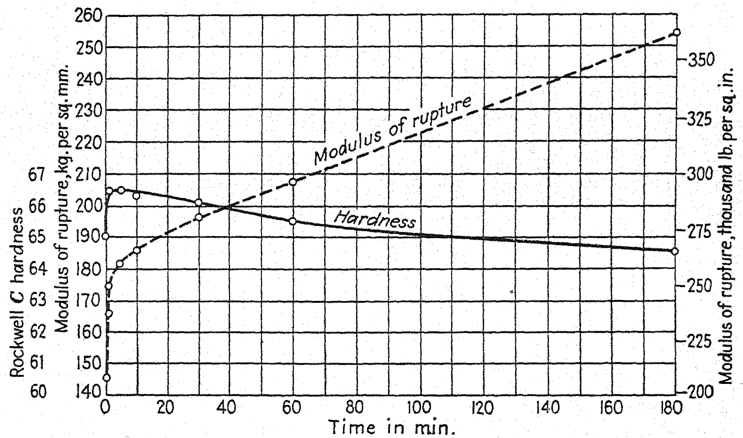

FIG. 80.—Effect of aging for 180 min. at 125°C. (255°F.) on the Rockwell *C* hardness and modulus of rupture of water-quenched high-carbon steel containing 0.98 per cent carbon and 0.43 per cent manganese. (*Steinberg and Subow.*[493])

lengths of time at room temperature, 75, 100, and 125°C. (165, 210, and 255°F.). The results are shown in Figs. 77 to 80. After holding the quenched specimens at room temperature for 5 months, the Rockwell *C* hardness had increased from 65 to 67 and the modulus of rupture from 210,000 to 230,000 lb. per sq. in. As the aging temperature was increased, the rate of hardening increased; for temperatures of 100 and 125°C. (210 and 255°F.) it increased to a maximum and then decreased. The modulus of rupture remained practically constant with time for an aging temperature of 75°C. (165°F.) but increased rapidly with time for aging temperatures of 100 and 125°C.

* Modulus of rupture (Biegefestigkeit) was determined, after Charpy impact values were found to be unreliable, by bending a 7 × 60-mm. (0.28 × 2.36-in.) round bar in a Brinell machine.

E. AUTHOR'S SUMMARY

1. When steel sections of large mass are quenched, the cooling rate decreases as the distance from the surface increases; as a consequence of this, the strength and hardness of the steel decrease progressively from the surface to the center, but usually not linearly. As engineers and designers who use large sections of wrought steel are interested primarily in the average, or possibly the minimum, properties of the piece, and as it is usually impracticable to test whole large sections in the laboratory, a specific knowledge of the effect of mass on the properties of steels of various carbon contents, heat treated in various ways, is of great importance.

2. The tensile and yield strengths of low-carbon steels containing 0.12 to 0.20 per cent carbon decrease as their size increases from 0.5 to 2 in.; the decrease in tensile strength with increasing section is greater with the higher carbon material. The elongation and reduction of area apparently increase slightly with increasing section. In the data quoted for these low-carbon steels there was a peak in the Izod impact resistance with sections of 1 to 1.5 in.

3. In 0.40 to 0.50 per cent carbon (1045) steel, quenched and tempered at 540 and 650°C. (1000 and 1200°F.), the tensile and yield strengths decrease markedly and the reduction of area decreases slightly with increasing section size. The elongation increases. The effect of mass is less pronounced when the material is oil quenched and tempered than when it is water quenched and tempered; and for the same quenching treatment the effect of mass is less pronounced when the steel is tempered at 650°C. (1200°F.) than when it is tempered at 540°C. (1000°F.).

4. There is such a wide variation in the data on 1045 steel as given by property-mass charts published in American handbooks and the data on a steel containing 0.45 per cent carbon and 0.78 per cent manganese as published by the (British) Steel Research Committee that the errors in one or the other should be corrected by further work.

5. In forgings, 25 to 57 in. in diameter, containing 0.17 and 0.30 per cent carbon, the tensile- and yield-strength values, in general, decrease from surface to core. In the annealed forgings the decrease varies from 10 to 20 per cent; in air- or oil-quenched

and tempered forgings of the same size and composition the decrease is less. There is no definite trend in elongation, reduction of area, or impact resistance.

6. In forgings, 25 to 75 in. in diameter, containing 0.17 and 0.30 per cent carbon, transverse properties are inferior to longitudinal properties, but, except in the case of impact resistance, the inferiority is not marked. Oil or air quenching followed by tempering at about 625°C. (1160°F.) improves the impact resistance of both longitudinal and transverse specimens throughout the section, even in forgings 57 in. in diameter. Forging with a reduction of 80 per cent has no appreciable effect on the properties throughout the section as compared with a reduction of 50 per cent.

7. The effect of mass on the torsional strength of water-quenched and tempered medium-carbon steel is about the same as on the tensile strength: (*a*) increasing the cross-section reduces the torsional strength, and (*b*) increasing the tempering temperature makes the mass effect less pronounced.

8. According to data presented, increasing the size of the section in a water-quenched and tempered 0.45 per cent carbon steel reduces the yield ratio. The investigators who published these data have suggested that the yield ratio may thus afford an excellent measure of the deep-hardening properties of a steel.

9. Average properties for large locomotive forgings of 0.45 to 0.50 per cent carbon steel, normalized and annealed, indicate that sections over 10 in. in the smallest dimension have about the same tensile and yield strengths but about 5 per cent lower elongation and reduction of area than sections less than 9 in. in cross-section. Spheroidizing large forgings lowers tensile and yield strengths and increases elongation and reduction of area.

10. The tensile properties of heat-treated high-carbon steels are of minor interest as compared with the properties by which one may judge the fitness of the steel for tools, dies, and the like; hence there are practically no data available on the effect of mass on the tensile properties of these steels. Moreover, the response of high-carbon steels to heat treatment differs widely, even if the steels are of the same composition (as determined by the usual routine chemical analysis). The chief factors responsible for this variable hardenability are apparently oxides and grain size. Data (Fig. 76) on the hardness across the section of eight 1-in.

bars of 1 per cent carbon tool steel illustrate this difference in hardenability. Only one of the eight steels hardened completely throughout the section.

11. Steels of the same chemical composition (as usually determined) exhibit different tendencies toward grain growth above the critical range. If one steel has a tendency to grow large grains at a certain temperature and another steel, of the same composition, has not, the two steels, after heating to this temperature and cooling, will have different grain sizes and different properties. Data are presented which indicate that: (*a*) the fine-grained steels have a slightly lower tensile strength and the same or slightly lower yield strength and 10 to 25 per cent higher elongation and reduction of area than the coarse-grained steels, and (*b*) the impact resistance at normal temperature of a steel with a grain size of 5 to 7 is two to six times the impact resistance of a corresponding steel with a grain size of 2 or 3.

12. Quenched high-carbon steels show a gradual increase in hardness and modulus of rupture upon aging at room temperature for periods up to 5 months. These changes are accelerated if the material is aged at temperatures of 100 to 125°C. (210 to 255°F.). There are not enough data on the magnitude of the changes in aging high-carbon steels at low temperature to be evaluated with certainty.

CHAPTER VIII

MANUFACTURE AND SPECIAL PROPERTIES OF GRAY AND WHITE CAST IRON

Composition and Constitution of Cast Iron—Production of Iron Castings—Special Properties of Gray Cast Iron—Author's Summary

The commercial iron-carbon alloys known by the generic term *cast iron*, including gray iron, white iron, and malleablized cast iron, are impure alloys of iron, carbon, and silicon. Notwithstanding the fact that cast irons are at least ternary alloys, they may be considered in this monograph—as they are considered by many in industry—as binary iron-carbon alloys with a variable small amount of silicon tacitly assumed to be present. This classification is justified by the universal custom of defining alloy cast irons as those containing nickel, chromium, vanadium, copper, molybdenum, or some other element intentionally added to produce an iron with special properties. Ordinarily a cast iron containing silicon is looked upon as an alloy iron only when enough has been added (about 15 per cent) to make the material corrosion resistant.* Although the base-line for a study of constitution of alloy cast irons should be the iron-carbon-silicon system, in a review and correlation of the properties it would be justifiable to compare the properties of alloy cast irons as given in other monographs of this series with the properties of silicon-containing iron-carbon alloys as given in this and the next two chapters.

The literature on cast iron is voluminous. Mellor[551] cited seven or eight hundred references to articles and books on cast iron. In preparing this discussion, a comprehensive survey of the literature on cast iron was not attempted,† and many of the

* High-silicon corrosion-resistant iron can be called a cast iron only because it is usually melted in a cupola or air furnace and cast in molds. The carbon content is much lower than is usual in cast iron.

† The discussion of the properties of plain cast iron as given in this and the next two chapters should be complete enough to serve as a satisfactory

references cited may be less important than others not cited. For more detailed treatments of cast iron reference should be made to books by Hatfield[258] and Piwowarsky[330] dealing chiefly with gray iron, those by Moldenke[396] and Hurst[308] which deal especially with melting practice, and those by Schwartz[94] and Schüz and Stotz[410] on malleable cast iron. From the historical point of view Keep's[6] book deserves mention.

A Symposium on Malleable Iron Castings,[429] published in 1931, and a Symposium on Cast Iron,[582] published in 1933, give excellent treatments of the respective subjects and were freely used in preparing the discussion given below.

A. COMPOSITION AND CONSTITUTION OF CAST IRON

A satisfactory definition of cast iron has not yet been formulated. This has been discussed in Chapter I where the various definitions proposed for the commercial iron-carbon alloys are given. As stated there, the dividing line between carbon steel and cast iron was arbitrarily placed, for the purpose of this monograph, at approximately 1.7 per cent carbon (point *E* of the iron-cementite diagram). The properties of alloys containing less carbon than this are included in the chapters on carbon steels; the properties of alloys with higher carbon are included in this and the next two chapters. Malleable cast iron may be considered as a species of cast iron. The malleability is not present in the as-cast, white iron but is produced by heat treatment. The term "malleablized iron" would be more exact than the one usually bestowed on this material.

90. Composition of Cast Iron.—Unalloyed cast iron usually contains from 2 to 4 per cent carbon, from 0.25 to 3 per cent silicon, and varying percentages (usually small) of manganese, sulphur, and phosphorus.* White cast iron and malleablized

---

"base-line" with which the properties of any alloy cast iron may be compared. A complete summary of the constitution and properties of plain and alloy cast irons, including a more comprehensive correlation of data than is made in the present monograph, will be attempted in a forthcoming monograph entitled "Cast Iron."

* Unalloyed cast iron may also contain small amounts of other impurities, picked up from the scrap. For example, Subcommittee XV of Committee A-3, A.S.T.M., on Impact Testing[586] reported one specimen of iron which contained 0.12 per cent tin, apparently the result of melting a charge in the electric furnace which contained tinned, or insufficiently detinned, scrap.

iron contain about 1 per cent or less silicon, and soft gray iron, cast in small sections, may contain as much as 3 per cent silicon.

In foundry practice, the hardness of the material produced is controlled by regulating the silicon and carbon contents. As either the silicon or carbon is increased, the amount of combined carbon decreases, so that low-silicon, low-carbon irons tend to be "white"—*i.e.*, they have a white fracture because the carbon is in the form of cementite rather than graphite; high-silicon, high-carbon irons, on the contrary, tend to be gray—*i.e.*, they have a gray fracture because the carbon is principally in the form of graphite. The influence of carbon and silicon on the character of the iron produced is discussed later.

The structure and properties of cast iron depend upon the melting and casting conditions as well as upon composition. An iron which is gray when cast in heavy sections may be mottled (a mixture of gray and white) or white when cast into smaller sections which cool more rapidly. Where one surface of a casting must be hard while the body of the casting must be soft, a block of metal, known as a "chill," is placed in the mold, so that it forms the wall for the face that is to be hard. This chill causes the metal adjacent to it to cool rapidly; this prevents, to a degree at least, the formation of graphite and results in a white fracture and a hard iron, while the other portions of the casting cool more slowly, which promotes the formation of graphite and results in a soft iron with a gray fracture. In such materials as rolls, the composition of the iron is adjusted and the cooling controlled so that the surface of the casting will be white, hard iron, while the interior is soft and gray, thus combining a wear-resistant surface with a fairly ductile core. The composition of white iron which is to be malleablized is regulated so that no graphite will form on casting, yet so that practically all of the cementite will be converted into graphite by an annealing treatment.

Because the structure and properties of an iron depend not only on the composition but on the size of section and on the melting and casting technique, it is impossible to prepare a table of analyses which will result in certain definite properties. Analyses of irons suitable for certain applications can, however, be chosen. Table 61 is a list of such analyses prepared by a committee of the American Foundrymen's Association.[503] An earlier table was given by Moldenke.[61]

91. Constitution of Cast Iron.*—The three major constituents of cast iron are alpha iron (ferrite), iron carbide (cementite or "combined carbon"), and graphite. With the exception of austenitic cast irons, which are highly alloyed and, therefore, not considered here, all cast irons contain alpha iron. This

TABLE 61.—RECOMMENDED ANALYSES OF PLAIN CAST IRONS FOR COMMON APPLICATIONS*

Composition, per cent					Typical charge, per cent			Use
Total C	Si	Mn	P	S	Pig	Return scrap	Steel scrap	
3.25	2.25	0.65	0.15	0.10	40	40	20	Automobile cylinders
3.35	2.25	0.65	0.15	0.10	40	40	20	Automobile pistons
3.40	2.60	0.65	0.30	0.10	50	45	5	Automobile castings, general
3.50	2.90	0.65	0.50	0.06	60	40	....	Automobile piston rings—individually cast
3.40	2.10	0.60	0.50	0.10	35	55	10	Agricultural implements—medium sections
3.50	2.20	0.55	0.70	0.10	40	60	....	Agricultural implements—light sections
3.25	2.25	0.50	0.50	0.09	50	50	....	Machinery—sections not over 1 in.
3.25	1.75	0.50	0.50	0.10	50	40	10	Machinery—1.5-in. sections
3.25	1.25	0.50	0.50	0.10	50	25	25	Machinery—2-in. sections
3.25	1.25	0.65	0.20	0.10	50	25	25	Pressure castings—air cylinders
3.40	1.75	0.80	0.35	0.09	55	20	25	Gas-engine cylinders—light
3.40	1.50	0.80	0.35	0.09	55	20	25	Gas-engine cylinders—medium
3.40	1.25	0.80	0.35	0.09	55	20	25	Gas-engine cylinders—heavy
3.30	2.00	0.50	0.60	0.10	50	40	10	Valves and fittings
3.50	1.15	0.80	0.10	0.07	70	30	....	Firepots and kettles
3.50	1.00	0.90	0.20	0.07	90	..	10	Ingot molds
3.60	1.00	0.75	0.20	0.07	60	40	....	Pots for caustic soda
3.60	1.75	0.50	0.80	0.08	70	25	5	Light and medium sand-cast water pipe
3.40	1.40	0.50	0.80	0.08	60	25	15	Heavy sand-cast water pipe
3.40	1.40	0.50	0.80	0.08	70	30	....	Soil pipe
3.20	1.10	0.85	0.35	0.17	†	43	30	Wearing plates
3.35	0.65	0.60	0.35	0.12	12.5	80	7.5	Car wheels (0.90 per cent combined C)
3.60	1.25	0.55	0.40	0.10	45	40	15	Chilled plow iron
3.75	0.85	0.50	0.40	0.10	45	40	15	Plow mold boards
2.75	2.25	0.80	0.10	0.09	....	10	85	Base for high-strength cupola iron; 5 per cent ferroalloys to be added

* American Foundrymen's Association.(503)
† 25 per cent scrap car wheels, 2 per cent spiegeleisen.

* The constitution of cast iron, based on the iron-cementite and the iron-graphite diagrams, is discussed in detail in Volume I, Chapters II to V, of this monograph. The constitution of cast iron, based on the iron-carbon-silicon system, is discussed in Chapters VII and VIII of "The Alloys of Iron and Silicon."(618)

constituent is not pure but is an iron-rich solution containing practically all of the silicon and at least a portion of the other elements found in small amounts in the iron. The appearance of the fracture and the hardness of the iron depend upon the relative amounts of the three major constituents; thus, in soft gray irons a large part of the total carbon is in the form of graphite with the remainder, sometimes less than 0.30 per cent, in the form of iron carbide. Mottled irons contain less graphite and more cementite—frequently about equal parts of graphite and combined carbon—and white irons contain practically all the carbon as cementite with little or no graphite present. Malleable iron made by the usual American practice contains practically no iron carbide; this constituent has decomposed on annealing into iron and nodular graphite (temper carbon).* The properties of cast iron depend primarily on the relative amounts of iron carbide and graphite and on the distribution of these two constituents in the matrix. Cast iron may also contain non-metallic inclusions, chiefly manganese sulphide. Because of the high carbon and silicon contents in cast iron the solubility of phosphorus in the matrix is limited; many irons contain enough phosphorus to form iron phosphide as a separate phase. The phosphide appears as part of a eutectic which consists of iron phosphide, ferrite, and, in some cases, cementite. This eutectic is known as *steadite;* 1 per cent of phosphorus by weight will produce approximately 10 per cent of steadite by volume.(582) Steadite being hard and brittle tends to increase the hardness and brittleness of iron.

In addition to the three major constituents of cast iron noted before, it is customary also to consider pearlite, the aggregate of ferrite and cementite, as a separate constituent. If this is done, the important constituents of commercial cast iron are *free ferrite*, *pearlite*, *free cementite*, *graphite*, and possibly *steadite*. In some grades of cast iron, and especially certain irons which are cast in thin sections and, consequently, cooled rapidly, the aggregate of ferrite and cementite, which in slowly cooled sections

* Malleable iron made by the "short-cycle" anneal is likely to contain some undecomposed iron carbide in a pearlitic matrix. The essential difference between this variety of malleable and the pearlitic high-strength cast irons (see page 300) is that in the former the free carbon is present as temper carbon and in the latter as flaky graphite.

TABLE 62.—STRUCTURAL COMPOSITION OF SOME TYPICAL IRONS*

Type, number	Fracture	Chemical composition, per cent							Structural composition, per cent by volume						Calculated specific gravity
		Graphite	Combined carbon	Si	Mn	P	S	Fe	Silicoferrite	Pearlite	Manganese sulphide	Steadite	Graphite	Free cementite	
1	Gray	2.10	0.80	1.10	0.75	0.20	0.07	94.98	6.36	84.59	0.35	2.00	6.70	.....	7.34
2	Gray	2.50	0.70	1.80	0.80	0.32	0.10	93.78	15.65	72.84	0.49	3.15	7.87	.....	7.23
3	Gray	2.65	0.65	1.25	1.00	0.10	0.12	94.23	21.98	67.89	0.60	0.98	8.35	.....	7.25
4	Gray	2.50	0.72	1.50	0.90	0.30	0.11	93.97	12.98	75.64	0.54	2.96	7.88	.....	7.25
5	Gray	3.09	0.40	2.40	0.55	1.05	0.10	92.41	39.14	40.75	0.48	10.12	9.51	.....	7.08
6	White	....	3.30	0.60	0.52	0.50	0.15	94.93	2.23	49.13	0.67†	5.21	None	42.67	7.66
7	Mottled	1.50	1.80	0.92	0.36	0.22	0.13	95.07	3.30	72.71	0.45†	2.23	4.84	16.26	7.43

* Symposium on Cast Iron.(582)

† Numbers 6 and 7 also contained 0.09 and 0.21 per cent iron sulphide respectively.

would assume the characteristic lamellae of pearlite, may assume a form which can more properly be termed fine pearlite or even troostite.*

The silicon of cast iron exists almost entirely in solid solution in the ferrite. The remainder exists in solution in the iron carbide; this ignores, of course, silicon which may be present in entrapped sand or other silica. The distribution of silicon in iron-carbon-silicon alloys is discussed at length in "The Alloys of Iron and Silicon."(618)

Table 62, from a symposium,(582) gives the analyses of several cast irons and the volume percentages of the structural constituents. Table 63, from the same source, gives the properties of the constituents shown in Table 62.

The hardness of unalloyed cast irons increases as the cementite increases, but the strength does not necessarily increase with the hardness. The hardness of gray iron depends primarily on the amounts of the constituents, while its strength and ductility depend upon the distribution of the constituents, particularly upon the size and distribution of the graphite. Gray cast iron may be considered as high-silicon steel in which graphite flakes are embedded; the carbon content of the "steel" is represented by the amount of combined carbon in the iron. Malleable iron is effectively a steel of very low carbon content containing nodules or nests of graphite. As compared with gray iron, malleable iron is very ductile; it owes its ductility to the fact that the graphite is in the form of nodules rather than in the form of plates. Graphite which forms at a high temperature, either by decomposition of cementite soon after the metal freezes or possibly by precipitating from the melt, exists in the form of flakes or plates, while graphite formed at lower temperatures, say 900°C. (1650°F.), tends to form fine nodules. In malleable iron all of the graphite is formed on annealing. Micrographs of white and gray cast irons are shown in Volume I, Chapter V, and of malleable iron in Volume I, Chapter VIII.

B. PRODUCTION OF IRON CASTINGS

The base-material used for iron castings may be melted in any one of several radically different types of furnaces and poured

* In chilled alloy irons it is even possible for martensite to be present in the microstructure.

into any one of several kinds of molds. The properties of the casting depend upon the conditions existing in the melting furnace and upon the kind of mold into which the metal is cast, as well as upon the composition of the materials charged into the furnace

TABLE 63.—PROPERTIES OF CONSTITUENTS GIVEN IN TABLE 62*

Constituent	Specific gravity	Approximate			Remarks
		Tensile strength, lb. per sq. in.	Elongation in 2 in., per cent	Brinell hardness	
Ferrite	7.86	50,000	40	95	Iron
Iron silicide† (FeSi)	6.17	Low	None	.....	1 per cent silicon forms 3 per cent silicide
Pearlite	7.846	120,000‡	15	240	A laminated structure containing 6.5 parts soft iron or ferrite to 1 part cementite.
Cementite (Fe_3C)	7.66	5,000	None	550 or more	A compound of iron and carbon containing 6.67 per cent carbon
Steadite	7.32	Small	Brittle	Very hard	Phosphide of iron (Fe_3P) and saturated solution of Fe_3P in iron. This constituent is 10 times (by volume) the amount of phosphorus (by weight)
Manganese sulphide	4.00	None	Brittle	.....	1.73 parts manganese to 1 part sulphur
Iron sulphide	5.02	Low	Brittle	.....	1.75 parts iron to 1 part sulphur
Graphite	2.55	None	Non-coherent		

* Symposium on Cast Iron.[582]

† Exists largely in solid solution in the ferrite-rich portion. In normal irons it is not discernible as a separate component.

‡ According to Belaiew,[357] the tensile strength of pearlite may be as high as 150,000 lb. per sq. in.

and the composition of the metal poured into the mold. The several common methods of producing iron castings are described on the next pages, but the influence of methods of production on the properties of the castings is treated later.

92. Melting Stock.—Most foundries make their iron castings from cold pig iron and scrap cast iron* (see Table 61, page 256).

* A few manufacturers, especially of large castings, use molten iron from the blast furnace.

Scrap steel may or may not be used in the charge, and with special production methods, which are not described here, cast iron may be made from a charge consisting wholly of steel scrap with suitable ferroalloy additions, and with, of course, the addition of carbon from the coke used in the cupola, or from a separate addition in the case of other furnaces. In earlier days the misnomer "semisteel" was applied to iron made from a charge containing steel scrap, used to reduce the carbon content. These lower carbon irons of higher strength are now classified among the "high-test" or "high-strength" irons with no reference to the method by which a composition suitable for the attainment of high strength is obtained. In a study of impact properties of cast iron by a Committee[586] of the American Society for Testing Materials, specimens were tested from heats in which varying amounts of steel scrap had been used, ranging from those containing no steel to those which contained as much as 85 per cent.

The iron scrap used may consist entirely of that produced in the foundry (gates, risers, defective castings, and the like), or it may contain a relatively large amount of purchased scrap. The composition of the charge must be regulated in accordance with the melting technique, as well as in accordance with the desired composition of the iron, for the analysis of the material usually changes during melting. As an example of this, Schwartz[94] found that in melting malleable iron in an air furnace the loss of certain elements as based on quantities present in the charge amounts to the following:

Element	Loss, Per Cent
Carbon	15.8
Silicon	31.4
Manganese	48.1
Phosphorus	0
Sulphur	−22.2 (gain)
Iron	1.2

Lorig, in a private communication, stated that the loss in silicon as found by Schwartz does not always occur; on the contrary, once the bath is hot, silicon frequently increases. Lorig supplied several illustrations in which the average pick-up in silicon, from the beginning of the heat to the end, was 0.02 to

0.11 per cent—for instance, from 0.90 to 0.92 per cent, and from 0.84 to 0.95 per cent silicon.

Pig iron is classified in three ways: (1) according to the method of manufacture, thus: coke pig, charcoal pig, and anthracite pig; (2) according to the use to which it is to be put, thus: Bessemer pig for steel making by the acid Bessemer process, basic pig for steel making by the basic open-hearth process, malleable pig for castings to be malleablized, foundry pig for ordinary iron castings, and forge pig, an inferior grade used in puddling and sometimes for foundry work; and (3) by chemical composition.

A number of years ago, when pig iron was sand cast, foundry iron was graded by breaking the pig and examining the fracture. From the size of the grains and the color of the fracture the amount of silicon in a slowly cooled pig could be estimated with fair accuracy. According to the fracture, iron was graded into

Table 64.—Classification of Principal Grades of Pig Iron*

Grade of iron	Silicon, per cent	Sulphur, per cent	Phosphorus, per cent
No. 1 soft (Southern) or No. 1 foundry (Northern)	2.75 to 3.25	Under 0.05	†
No. 1 foundry (Southern) or No. 2*X* (Northern)	2.25 to 2.75	Under 0.05	†
No. 2 foundry (Southern and Northern)	1.75 to 2.25	Under 0.05	†
No. 3 foundry (Southern and Northern)	1.25 to 1.75	Under 0.05	†
Gray forge	Under 1.75	Under 0.075	†
Malleable	0.75 to 2.00	Under 0.050	Under 0.2
Bessemer	1.00 to 2.00	Under 0.050	Under 0.1
Low phosphorus	1.00 to 2.50	Under 0.040	Under 0.04
Basic	Under 1.25	Under 0.050	†
Basic Bessemer (Europe)	Under 1.00	Under 0.050	2.0 to 3.0

* Tiemann.(606)

† Phosphorus limits vary in different parts of the country, usually from 0.4 to 1.1 per cent. Manganese varies, usually it is less than 1.0 per cent, excepting basic Bessemer which may contain up to 2.0 per cent.

No. 1 foundry or No. 1 soft, No. 1 foundry or No. 2*X*, No. 2 and No. 3 foundry, and gray forge. Compositions of these are given in Table 64. These classifications have varied in different parts of this country and abroad, causing at times considerable

confusion. The adoption of machine casting in many plants, with its rapid chilling which changed the appearance of the fracture, forced the use of chemical analysis for grading foundry and other irons, but the old classifications are still widely used. The grading and the chemical composition of the various grades of foundry, steel-making, and other irons are discussed by Boylston,(251) Camp and Francis,(150) and in detail by Tiemann.(666) The composition limits of the chief grades, according to Tiemann, are as shown in Table 64. The total carbon content of all these irons is about 3.50 to 4.25 per cent.

Of the 31 million tons of pig iron produced in 1930 by the blast furnaces of this country, 59.4 per cent was used for the manufacture of basic open-hearth steel, 23.5 per cent for the manufacture of acid Bessemer steel, 11.6 per cent in the foundry, and 5.1 per cent for malleable castings. The rest, less than 0.5 per cent, was forge iron, and mottled and white irons.

93. Melting Practice.—The oldest form of melting furnace, the cupola, still produces the major part of the molten iron which is poured into castings. As is well known, the cupola is a shaft furnace into which the metallic materials, coke, and flux are charged at the top, and air is blown through tuyeres near the bottom. The combustion of the coke supplies the heat necessary to melt the iron, which collects on the hearth at the bottom from which it is drained continuously or at frequent intervals. During melting, part of the manganese and silicon in the charge is lost by oxidation, and sulphur is picked up from the fuel. The change in carbon content varies with the composition of the iron charged and with operating conditions of the cupola such as the nature of the coke, condition of the slag, and other variables, most of which are under the control of the operator. Carbon control in the cupola has been discussed recently by Langebeck.(806) After the material is charged the only control of the composition of the iron is by means of the blast. Experienced operators, however, produce iron of very uniform composition. The many factors which influence the composition and character of cupola-melted iron have been recently studied by Neustätter;(645) it has not been considered necessary to review these here. Judson(536) has described the melting of irons of two compositions in two cupolas and their mixing in the ladle to produce the desired composition. This is stated to give properties not attainable

by the use of one cupola operated to produce a melt of the same composition as the mixture.

The air or reverberatory furnace may be used for producing high-grade cast iron. In this furnace the iron is melted on a hearth, and the composition of the iron can be controlled by means of additions to the bath while the charge is melting or after it is molten. Still a third type of furnace in which combustion is used as a source of heat is the rotary furnace equipped for heating with pulverized coal. This furnace has been used extensively in Germany and other European countries but is just coming into use in this country. Electric furnaces are used for producing some classes of high-grade cast iron. They may be used for superheating or for slightly refining metal melted by a cheaper process. This latter combination method is called "duplexing." In malleable practice, as described by Schwartz,[94] molten iron and molten steel from the Bessemer converter may be mixed to give the desired composition, which is followed by refining and superheating in an electric furnace; this is "triplexing."

94. Casting Gray Iron.—The practice used in gray-iron foundries for the preparation of loam, green-sand, and dry-sand molds into which the molten iron is poured has been described in so many elementary textbooks and is so well known that further reference to this practice in this monograph is unnecessary. In addition to pouring molten iron into sand molds, gray-iron castings are also produced in stationary permanent metal molds, and some pipe and other tubular products are now cast into revolving molds; these are the familiar centrifugal castings. In general it is not customary to use hot molds for gray-iron castings; in a recent development, however, high-strength pearlitic iron is produced by pouring low-silicon, low-carbon iron into a preheated mold.

The change of volume of iron-carbon alloys on melting is discussed in the Symposium on Cast Iron.[582] Sauerwald[235] found that a white cast iron (4.15 per cent carbon, 2.2 per cent manganese) increased 1.4 per cent in volume on melting while gray iron (3.32 per cent carbon, 2.76 per cent silicon, 0.49 per cent phosphorus) decreased 0.6 per cent. The abnormal decrease in the gray iron was ascribed to the volume contraction resulting from the formation of cementite. Bardenheuer and Ebbefeld[148]

attributed the expansion of gray iron during solidification to gas in the metal. Bardenheuer and Bottenberg[434] continued the investigation with particular attention to the factors which control the amount of gas and noted that the gas evolution depended upon the melting conditions as well as the composition. They also ascribed part of the expansion of gray iron on solidification to the formation of graphite.

In general, it would appear that iron-carbon alloys expand on melting and contract on solidification. During solidification the evolution of gas and the formation of graphite may lead to an increase in volume.

95. Fluidity and Shrinkage.—One of the most important characteristics of cast iron is its fluidity, *i.e.*, its ability to fill a mold and to reproduce accurately the design of the pattern. The ability of filling minute impressions in a mold must be dependent, to some extent at least, on the viscosity and surface tension of the liquid iron. Data on viscosity are not very reliable; on surface tension, as noted by Piwowarsky[330] and Krynitsky,[629] trustworthy values are lacking. Thielmann and Wimmer[241] and Esser, Greis, and Bungardt[696] among others determined viscosity. The former used very low silicon pig iron which solidified as white iron; the latter used irons which were gray after solidification. Comparison of their results[696] indicated that gray iron had a higher viscosity than iron which solidified white. Although the two groups of irons were tested at the same laboratory, the experiments were not made at the same time or by exactly the same procedure; hence, the conclusion that gray iron has in general a higher viscosity than white iron should be verified by further work. Piwowarsky,[330] in reviewing the work of Thielmann and Wimmer, concluded that viscosity* did not afford a direct measure of fluidity† and that surface tension was probably more important than viscosity. As noted above, Piwowarsky commented on the lack of reliable data for the surface tension of cast irons.

* It should be noted that the terms viscosity and fluidity of the foundryman are not reciprocal, as in physics.

† In commenting upon this portion of the manuscript, J. T. MacKenzie wrote: " . . . the property known by the technical term 'viscosity' is in itself of small concern in cast iron. . . . The actual viscosity of the material has very little to do with its fluidity (running ability) in the foundry aside from that due to temperature only."

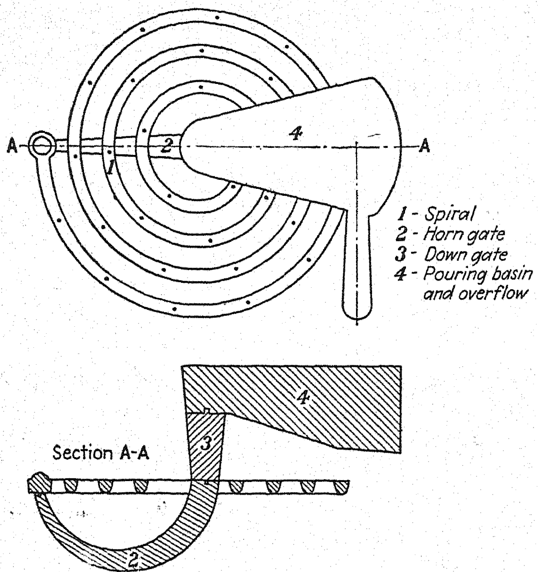

FIG. 81.—Pattern assembly used for determining fluidity of cast iron. (*Saeger and Krynitsky*.[486])

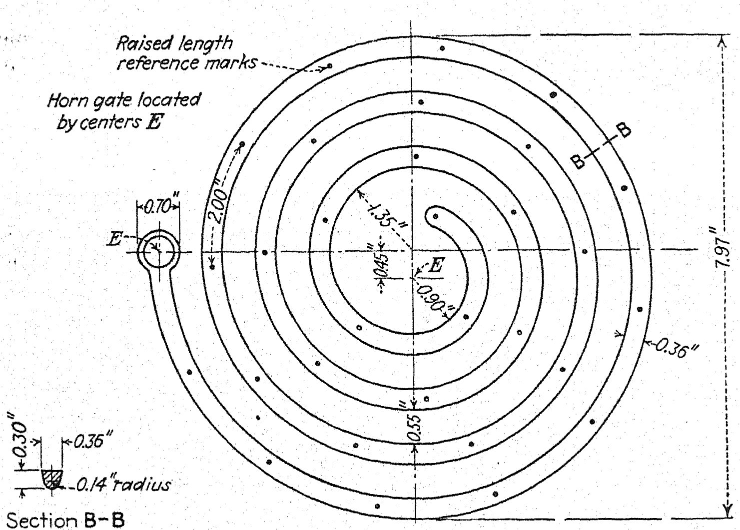

FIG. 82.—Dimensions of the spiral shown in Fig. 81. (*Saeger and Krynitsky*.[486])

A direct method for determining the fluidity of cast iron and one which gives evidence of becoming standardized consists in pouring molten iron into a spiral which is so long and thin that the molten metal does not fill the mold completely. The length of the spiral produced is taken as a measure of the fluidity of the molten iron. A spiral and a method of gating found to be satisfactory by Saeger and Krynitsky[486] are shown in Figs.

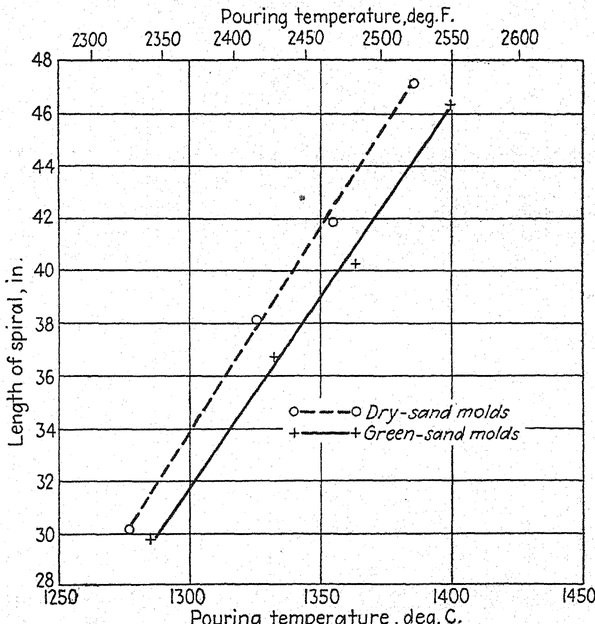

FIG. 83.—Fluidity of cast iron containing 3.44 per cent carbon, 2.38 per cent silicon, and 0.47 per cent phosphorus, cast in green- and dry-sand molds. (*Saeger and Krynitsky.*[486])

81 and 82. The pouring basin is arranged with an overflow so that there is a constant head of molten metal during casting. Either a dry-sand or green-sand mold may be used, and the end of the spiral should be vented. It was found that consistent and reproducible results could be obtained with aluminum and cast iron. The lengths of spiral obtained from a cast iron cast at different temperatures in both types of mold are shown in Fig. 83. The iron contained 3.44 per cent carbon, 2.38 per cent silicon, and 0.47 per cent phosphorus. In discussion of the paper just mentioned, Spencer gave some data on fluidity, as

indicated by casting spirals, which showed that it was a linear function of the sum of the carbon content plus 0.3 times the silicon content in the range 3.2 to 4.0 per cent carbon plus 0.3 times the silicon. The fluidity continues to increase, but not as a linear function of this sum, until the eutectic value of 4.2 or 4.3 per cent for the sum of carbon plus 0.3 times the silicon is reached, after which the fluidity decreases. Spencer's values were corrected to give fluidity values for a fixed temperature above the liquidus.

The fluidity of three cast irons, as dependent upon composition and maximum temperature to which the irons were heated before pouring, was reported by Saeger and Ash.[653] The compositions of the irons were as follows:

Iron	Composition, per cent						
	Total carbon	Graphite	Combined carbon	Si	Mn	S	P
A	3.55	3.15	0.40	2.73	0.24	0.021	0.82
B	Mixture of 80 per cent of *A* and 20 per cent open-hearth ingot iron						
C	3.79	3.13	0.66	1.32	0.73	0.06	0.12

The spiral shown in Fig. 82 was used. Results are given in Fig. 84. These data show that iron *A* had the highest fluidity. Peculiarly, although the phosphorus and silicon of iron *B* were much higher than in iron *C*, the fluidity of the latter was much better. Regarding the effect of heating to high temperatures before pouring Saeger and Ash stated:

> The running qualities of these irons are dependent upon the liquidus temperature, and apparently are not influenced by heating temperatures above the pouring temperature. For a given pouring temperature, the irons having the lower liquidus temperatures have superior running qualities.

In a recent paper, Périn[647] reported the results of a number of tests on the fluidity of cast iron as determined by pouring spirals. The effects of different amounts of carbon, phosphorus, and silicon on the fluidity of iron were determined. In binary iron-carbon alloys, maximum fluidity was found in alloys of a eutectic composition (4.3 per cent carbon).

For white iron for malleablizing, Toúceda[818] found another type of fluidity tester more suitable than the spiral tester. The test is used primarily for control work, to make sure that the metal itself is not responsible for mis-runs. Toúceda poured round and square fluidity bars, ¼ × ¼ × 12 in. and ⅛ × ⅛ × 12 in., horizontally from a pouring basin and sprue. The

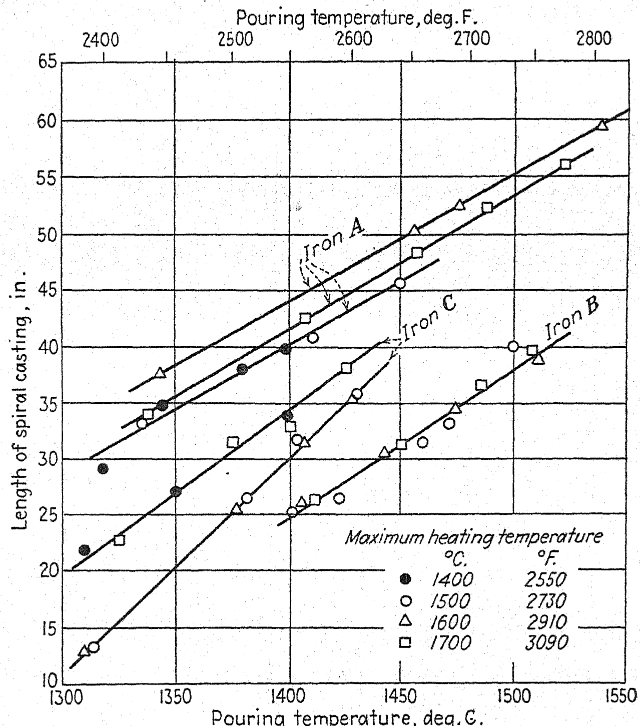

FIG. 84.—Effect of maximum heating temperature on the fluidity of cast iron. See table on page 268 for composition. (*Saeger and Ash.*[653])

patterns for the square bars were set on an edge in the mold so that the bar was diamond shaped as poured. With a pouring temperature of 1565°C. (2850°F.), iron which will run satisfactorily in the usual malleable casting will completely fill the ¼-in. molds and more than 8 in. of the ⅛-in. molds.

Volume changes occurring on casting have been studied by many investigators, including Saeger and Ash,[407,506,564,565,653] Wüst and Schitzkowski,[96] Bardenheuer and Ebbefeld,[148] and Bardenheuer and Bottenberg.[434] In addition to their

experimental work, most of these workers reviewed the literature on the subject. Saeger and Ash considered volume changes in the metal prior to solidification, volume changes during solidification, volume changes during cooling from the solidification temperature, and the effect of heating to temperatures above the pouring temperature. They determined the specific volume of liquid irons at different temperatures by sampling with a container of known capacity and determining the weight of the sample. They estimated the volume change in the cooling

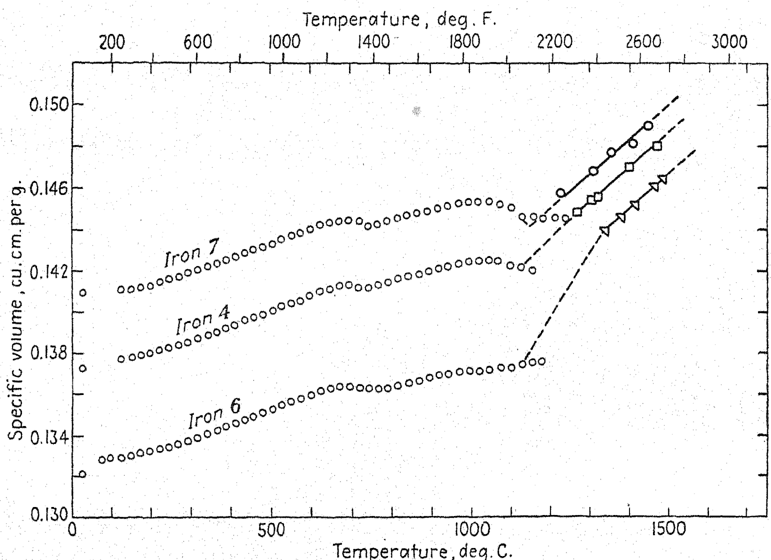

FIG. 85.—Specific-volume versus temperature curves for three cast irons. (*Saeger and Ash.*[565])

solid from the change in length of a rod as it cooled from the solidification temperature to ordinary temperature. From the density of the iron at ordinary temperature and the change in volume on cooling from the solidification temperature the specific volume immediately after solidification was calculated. Data were then available for calculating the specific volume at any temperature and for determining the volume change through any temperature range. Curves for three cast irons showing the volume change occurring on cooling from the liquid state to ordinary temperature are shown in Fig. 85. The analyses of these irons as cast into 1.5-in. sq. bars were:

Iron number	Composition, per cent					
	Total carbon	Graphite	Si	Mn	S	P
7	3.67	3.26	2.10	0.54	0.05	0.46
4	3.10	2.31	1.69	0.48	0.04	0.35
6	2.18	0.56	1.24	0.35	0.04	0.27

The increase in volume on cooling from just below the freezing point, observed with irons 7 and 4, may be due in part to the formation of graphite, but most of the expansion is probably

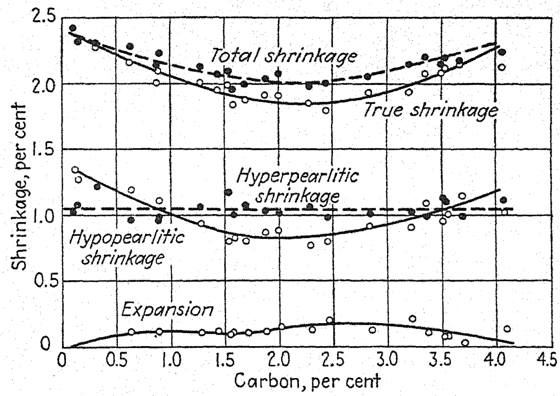

FIG. 86.—Effect of carbon on the casting shrinkage of cast iron. (*Wüst and Schitzkowski, according to Piwowarsky.*[751])

caused by the liberation of dissolved gases.[434] Schwartz* stated that the net shrinkage of gray iron is about the same as the shrinkage of white iron less the expansion produced by liberating an equivalent amount of graphite in malleable practice.

The effect of carbon content on the casting shrinkage of a series of iron-carbon alloys is shown in Fig. 86, taken from Piwowarsky[751] who ascribed it to Wüst and Schitzkowski.

96. Alloying Gray Cast Iron.—The mechanical properties of gray cast iron, with its matrix interspersed with graphite flakes, cannot be altered so much by alloying as is the case with steel, since the discontinuities limit the strength which can be obtained. In pearlitic irons the properties of the pearlitic cementite can

* Private communication.

be altered by the introduction of carbide-forming elements, particularly chromium and molybdenum. The more stable, complex carbides are generally harder than cementite and help to produce abrasion resistance as well as to make the iron more stable against growth. It is often found that such additions render the iron too liable to chill so that a graphitizing element which will also dissolve in the ferrite and strengthen it, such as nickel or copper, has to be added to correct the chilling tendency. Elements which dissolve in ferrite may also be used alone for strengthening and toughening. Alloy cast iron is rather new but, since cast iron is similar to steel save for its graphite content, most or all of the alloying elements found useful in steel may similarly serve to improve the properties of cast iron.

97. White and Chilled Cast Irons.—White-iron castings are usually poured into sand molds; the carbon and silicon contents are adjusted to produce a completely white structure or a white outer layer of any desired thickness. In chilled iron, as mentioned on a previous page, the white portions of the casting are produced by inserting metal chills into the mold, which produce very rapid cooling of the molten iron in contact with them, thus preventing the formation of graphite in that part of the casting. In general, most cast irons will be white if cooled with sufficient rapidity, and it is, therefore, possible to obtain both white and gray iron in the same casting by controlling the cooling rates of the different parts, provided iron of the proper composition is used. In castings which are to be malleablized the iron is white throughout the entire cross-section.

White cast irons do not lend themselves to classification on the basis of properties, because the structure and its accompanying properties may vary widely from casting to casting and even from one part to another part of the same casting. Thus a casting may be entirely white, it may be mostly gray with only a thin outer layer of white iron, or it may be white in some portions, gray in others, and mottled in still others. Within limits, the hardness and brittleness of white irons may be controlled by regulating the carbon content; for example, irons with the highest carbon content are the hardest and most brittle.

Pohl[813] recently reported data on the effect of variations in the carbon and on the effect of temperature and time of heating before pouring on depth of chill. He found that, as the carbon

increased from 2.5 to 4.1 per cent, the depth of chill in specimens 130 × 40 mm. (5.12 × 1.41 in.) in cross-section and 150 mm. (5.91 in.) long decreased almost linearly from about 70 mm. (2.76 in.) to about 10 mm. (0.39 in.).

The effect of the presence of graphite in the charge was studied by melting a 3.85 per cent carbon iron in a crucible, using charges of all-white, all-gray, and equal parts of each. On heating to about 1290°C. (2355°F.) and casting at about 1215°C. (2220°F.), the all-gray charge gave a 12-mm. (0.47-in.) chill, the charge containing equal parts of white and gray iron a chill of 16 mm. (0.63 in.), and the all-white charge a chill of 20 mm. (0.79 in.). But when the melt was heated to about 1385°C. (2525°F.) and poured at 1250°C. (2280°F.) the depth of chill varied only from 29 to 33 mm. (1.14 to 1.30 in.).

The effect of temperature and time of heating was brought out by heating the same iron to 1220, 1355, and 1465°C. (2230, 2470, and 2670°F.) and pouring at 1185°C. (2165°F.); the depths of chill were respectively 18, 31, and 46 mm. (0.71, 1.22, and 1.81 in.).

On heating to 1400°C. (2550°F.) and pouring all specimens at 1230°C. (2245°F.), there was no chill when the melt was not held at 1400°C. (2550°F.) but was at once cooled and poured. Holding for 30 min. at 1400°C. (2550°F.) produced only a mottled surface, but holding at that temperature for 1 and 1½ hr. resulted in 1 and 12 mm. (0.04 and 0.47 in.) chill respectively. The core beneath the chill showed finer graphite and higher strength with the longer periods of heating.

A series of irons ranging from 2.9 to 4.1 per cent carbon was poured hot and cold, the hot pour being 90 to 220°C. (160 to 400°F.) higher than the cold. The scleroscope hardness of the chill was found to average around 1.5 points higher for the hot-poured specimens.

C. SPECIAL PROPERTIES OF GRAY CAST IRON

For many of the applications of cast iron, properties other than those included broadly under the classification mechanical properties are of great importance in determining the usefulness of the material. Among these are machinability, growth, and weldability. Others which may be mentioned briefly are damping and some of the physical characteristics such as thermal

expansion and thermal conductivity, and electric and magnetic properties. Data on these are outlined below.

98. Machinability.—Ordinary gray cast iron is generally considered to be an easily machinable material. Even a soft iron, however, may be encased in a very thin layer of chilled iron that renders machining difficult. Thin sections or corners of castings made from iron that does not show a strong enough graphitizing tendency may be white or mottled and, therefore, almost unmachinable. Burned-in sand on the surface of the castings may also render machining difficult.

Results reported by Wallichs and Dabringhaus[420] illustrate the difference of machinability of gray cast irons of different compositions and how machinability varies with strength and hardness. Seventeen irons were investigated whose tensile strength ranged from 17,000 to 50,000 lb. per sq. in. The specimens machined were in the form of tubes having an exterior diameter of 540 mm. (21 in.) and a wall thickness of 20 mm. (0.8 in.). Using similar lathe tools, the cutting speeds required to cause the tool to fail in 60 min. were determined for each iron. An attempt was made to determine the machinability of the surface as cast, but lack of soundness of the specimens made this impossible. The machinability values reported are, therefore, those for the body of the castings.

The data are shown in Figs. 87 and 88, in which the cutting speed for tool failure in 60 min. is plotted against tensile strength and hardness. Each plotting point represents the data for one cast iron. The iron whose value is shown by the solid circle was cast in a chill mold and then annealed. No relation could be found between composition and machinability or between structure and machinability. There was, however, a fairly good relation between tensile strength and machinability and a good relation between hardness and machinability. According to the data shown in Fig. 87, the machinability as measured by tool life is decreased more than 50 per cent as the strength is increased from 12 to 35 kg. per sq. mm. (17,000 to 50,000 lb. per sq. in.).

Data reported by Priester and Curran[231] indicated that the machinability of cast-iron pistons could be increased sixfold by annealing; the annealing decreased the strength and hardness from 30 to 40 per cent.

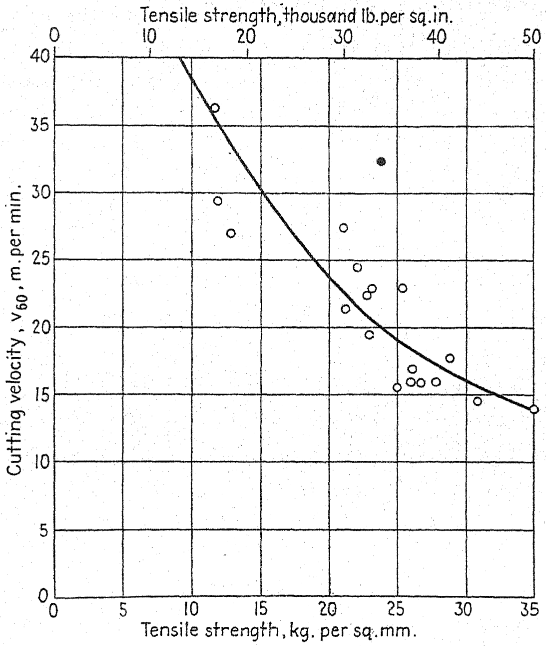

FIG. 87.—Relation between lathe cutting speed (for tool failure in 60 min.) and tensile strength. (*Wallichs and Dabringhaus.*[420])

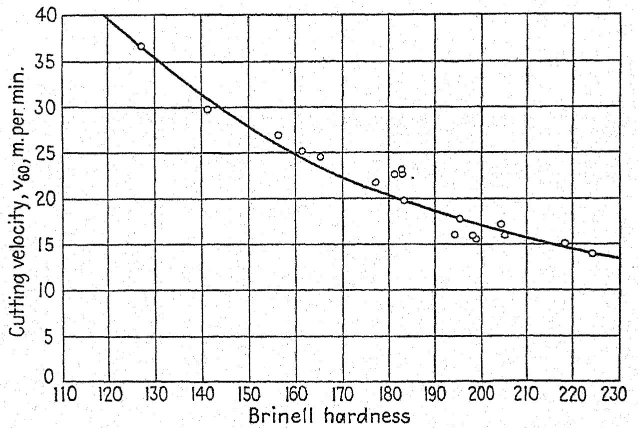

FIG. 88.—Relation between lathe cutting speed (for tool failure in 60 min.) and hardness. (*Wallichs and Dabringhaus.*[420])

Naturally, tests other than lathe tests have been used for determining the machinability of cast iron, the drilling test having been the favorite. In some cases surface finish may be quite important, and tool life apparently bears no direct relationship to the character of the machined surface. An interesting discussion of machinability of cast iron and references to some important studies will be found in the Symposium on Cast Iron.(582)

Diepschlag and Eggert(446) concluded that the machinability of cast iron increases with carbon content, while high silicon is not advantageous unless the carbon is over 4 per cent; with carbon below 3.5 per cent, high silicon was found to be detrimental. This comment relates especially to machining by chipping tools.

99. Wear Resistance.—Good summaries of the present-day understanding of the wear resistance of cast iron will be found in the Symposium on Cast Iron(582) and in a recent article by Boegehold.(593) Most of the discussion below is from the latter work.

Boegehold pointed out that of the many tests on wear resistance of cast iron no two were conducted under like conditions, and apparently no investigator has ever attempted to check the work of another. But in spite of this lack of agreement of testing methods, it is remarkable that so many of the conclusions agree. Methods of making wear tests may be grouped as follows:

1. Service tests.
2. Accelerated service tests.
3. Laboratory tests in which service conditions are simulated as nearly as possible.
4. Laboratory tests under arbitrary conditions.

Although a large number of service tests is continually in progress, results of these are rarely reported in the literature. Well-conducted service tests are naturally superior to the other types of tests in that they yield the most trustworthy data.

In an accelerated service test Bornstein(289) determined the wear in tractor-engine cylinders when run for 20 hr. in high gear with dirt or abrasive injected into the oil every 30 min. Alloy-iron cylinders having a Brinell hardness of approximately 150 showed much less wear than plain iron having a Brinell hardness in the neighborhood of 125.

Recent studies[723] in England have brought out the idea that what has usually been considered as abrasive wear in engine cylinders may be largely a matter of chemical corrosion by the products of combustion rather than a mechanical removal of metal.

A great variety of laboratory wear tests of both types mentioned on the previous page have been used, and most of the information found in the literature was obtained by some form of laboratory test. According to Boegehold:[593]

> Results obtained in laboratory tests can be relied upon only when materials of known value in service are first subjected to the laboratory test and are rated in the same order as determined in service. When this can be done, the laboratory test becomes of great value, because it permits the testing of a much greater volume of materials than can economically or conveniently be tested in service. The attainment of the test conditions which will give this result is sometimes very difficult. The logical course to pursue, obviously, is to make the test conditions as nearly like service conditions as is possible.

Söhnchen and Piwowarsky[660] also found that wear resistance varies according to type of wear but considered that, in general, pearlitic irons are more wear resistant than ferritic irons.

The following conclusions of different investigators have been summarized by Boegehold:[593]

1. From results obtained with the brake-shoe type of test by Lehman:[186]

a. Brinell hardness does not indicate the frictional-resistance quality.

b. Chemical composition cannot be used as a guide for judging abrasion qualities.

c. Abrasion resistance against steel is increased with increasing pearlite content and is influenced unfavorably by the presence of phosphide eutectic.

d. Gray cast iron with full pearlitic structure has the minimum abrasion when run against steel or cast iron.

2. From data obtained by Klingenstein,[382] using a brake-shoe test of Spindle type:

a. In annealed cast iron (ferritic) of below 160 Brinell hardness, the wear showed little change with phosphorus up to 0.55 per cent; with phosphorus increasing from 0.55 to 1.9 per cent, wear was reduced 33 per cent.

b. In pearlitic cast iron, not annealed, having Brinell-hardness values of 210 to 240, wear was very little affected by phosphorus up to 0.70 per cent. With phosphorus increasing from 0.70 to 1.35 per cent, wear was reduced 25 per cent; amounts above 1.35 per cent phosphorus had no further effect.

c. The pearlitic iron of 210 to 240 Brinell hardness wore only one-half as much as the annealed iron of 160 Brinell.

3. From tests with a blunt drill by Klingenstein:[466]

a. Minimum wear occurs when two parts are of equal hardness.

b. An increase in phosphorus from 0.4 to 1.0 per cent decreased wear 42 per cent.

c. An increase in chromium from 0.3 to 0.7 per cent decreased wear 25 per cent.

d. An increase in chromium plus nickel from 0 to 1.8 per cent decreased wear 25 per cent.

e. An increase in manganese from 1.0 to 1.65 per cent decreased wear 8 per cent.

f. An increase in silicon from 1.15 to 2.0 per cent increased wear 65 per cent.

4. From results obtained by Robin,[18] using an abrasive test with emery paper:

a. Resistance to wear increases with the amount of free cementite and with the percentage of phosphorus.

b. Free graphite does not appear to influence the results.

Boegehold himself drew the following general conclusions:

Certain conclusions to the contrary, the general picture obtained from all this work is that increased hardness, or the addition of elements which form hard constituents in the iron either by going into solution or by combining chemically, results in decreased wear. It is unfortunate that those who have published conclusions contrary to this picture have not given more details as to the microstructure and history of the metals tested or other details which might explain the deviation from a naturally expected result. The reason for exceptions to the general conclusions above is not now known, so that future work on wear testing should be conducted for the purpose of establishing these causes.

100. Growth.—The term "growth" when referred to cast iron applies to a permanent increase in dimensions which may occur when the iron is repeatedly heated and cooled or when it is held at high temperature. Only under corrosive conditions does notable growth occur at temperatures below 540°C. (1000°F.). The increase in volume due to growth may amount to as much as 50 per cent. A good review of the most important studies on the growth of cast iron was prepared by Piwowarsky and Esser.[279] Several articles by Donaldson[152,209,253] are mentioned under low-temperature annealing (page 336). A book by Piwowarsky[330] also contains a good summary of the literature on the subject. Bolton and Bornstein[437] gave an interesting discussion of growth and included a bibliography of 44 references.

Remmers[482] determined the causes of growth and concluded they were graphitization, oxidation along coarse graphite flakes, and mechanical swelling created by finely fracturing the slightly ductile structure of gray cast iron. Wood[672,673,674] also discussed the growth of plain irons and the effect of alloying elements upon it.

Work by Oberhoffer and Piwowarsky[167] indicated that a silicon-free alloy containing 4.3 per cent carbon had practically no tendency toward growth when repeatedly heated to temperatures which caused appreciable growth in normal cast iron. Sipp and coworkers[237,356] found that low carbon as well as low silicon tends to improve resistance to growth. Those who have studied the subject have found that the tendency toward growth increases directly with the silicon content or with the content of most other graphitizing elements. Conversely, those elements which retard graphitization tend to retard growth. This is equivalent to saying that carbide-forming elements, such as chromium, tend to suppress growth, while graphitizing elements, such as silicon and aluminum, tend to promote growth.

Growth of iron heated in an oxidizing atmosphere results chiefly from an oxidation or corrosion of the interior of the castings which is made possible by penetration of the atmosphere. Other processes, however, are active in bringing about growth, as is shown by the following conclusions drawn by Piwowarsky and Esser:[279]

1. An appreciable part of the increase in volume observed during the first stages of growth results from decomposition of the combined carbon.
2. Growth may be in part attributed to stresses introduced by thermal changes which cause minute fractures to form by the mechanisms suggested by Kikuta[88] and Benedicks and Löfquist.[205] Stresses can also result from the differences in expansion of the iron, graphite, and non-metallic impurities on change in temperature. After cracks have been formed the body of the casting can be rapidly oxidized, thereby causing a rapid increase in volume.
3. Gases entrapped in the casting can account for but a minute part of the growth. However, they may have a catalytic effect in regard to graphitization or other reactions. If entrapped gases are of an oxidizing nature, they act just as does the atmosphere in causing oxidation and growth.
4. A high density tends to retard growth as it hinders carbide decomposition and the entrance of gases.

5. Fine, temper-carbon-like graphite hinders the entrance of ambient gases and, therefore, tends to lessen growth. From the standpoint of the deterioration described under 2 this form of graphite would also tend to retard growth.

Growth at a temperature of 600°C. (1110°F.) was recently studied by Scheil,[568] who found that cementite was decomposed at this temperature and that an oxide of some sort, probably a silicate, formed in the outer layer of the iron. He found that the percentage increase in growth increased as the section decreased and pointed out that laboratory samples did not always give a correct idea of the amount of growth to be expected in commercial parts. Thyssen[779] has recently discussed the alloy "non-growth" cast irons developed by the British Cast Iron Research Association.

101. Hot Working.—Cast iron is seldom subjected to hot working, but in the austenitic state it will stand a small amount of hot deformation. This is utilized in making dies for deep drawing such objects as automobile fenders and for stamping out forks and spoons, by casting nearly to shape, heating into the austenite field—to about 980°C. (1800°F.)—and pressing in a master die which gives the working die its final shape.[562] In order to improve the wear resistance, chromium is often added[598] together with nickel to balance the chilling action of the chromium. The dies may be further improved by oil quenching and tempering.

102. Welding.—Among the restrictions in the use of cast iron as compared with other structural materials is the relative difficulty of welding. It is difficult to prepare welding rods or electrodes of the composition which will give a deposit of the same composition as the casting. The rapid cooling of the small volume of fused metal in contact with the body of the casting, as well as the rapid cooling of the metal about the weld which has dissolved graphite by being heated past the critical range, tends to produce a chilled area in and about the weld. Some details of precautions for preheating, welding, and annealing after welding are given in the Symposium on Cast Iron.[582]

Coatings of silicon carbide on welding rods have been suggested[337] the better to ensure soft welds; bronze, Monel metal and similar non-ferrous welding rods also find much use.

103. Damping.—A property of cast iron, favorable for many applications, but one not quantitatively studied till recently and whose evaluation is still poorly understood, is its damping power—the ability to absorb energy and stop vibration. Cast iron would make a poor tuning fork, for its absorption of energy causes the vibration to die out rapidly. The danger from building up of critical vibration which may cause a machine to shake itself to pieces or to build up a node of stress that might result in a fatigue failure is therefore much smaller with a cast-iron base for a machine, for example, than with a steel base. Figure 89 from Knackstedt[383] compares the damping power of some metals and of rubber. It will be seen that vibrations die out in cast iron at a rate more like that in rubber than in most metals. The testing conditions for the curves given in Fig. 89 were not strictly comparable, but the general behavior of the different materials is roughly indicated by the curves. Damping, and its relation to other properties, is discussed in more detail in Chapter XVII. The relation between damping and the endurance limit of cast iron is touched on briefly in Chapter XI, section 151.

104. Miscellaneous Physical Properties of Cast Iron.—Summaries of the thermal expansion, thermal conductivity, electric conductivity, and magnetic properties of cast iron are included in the Symposium on Cast Iron,[582] with selected bibliographic references to important articles on these properties which appeared prior to 1933.

According to this symposium, the coefficient of thermal expansion from room temperature to 1070°C. (1960°F.) varies from 0.0000092 to 0.0000169 per °C., increasing as the carbon decreases.

Sirovich and Vanzetti[773] studied the thermal expansion from room temperature to 400°C. (750°F.) of iron containing 3.0 to 3.6 per cent total carbon and 2.05 to 2.35 per cent silicon, with sulphur from 0.07 to 0.15 per cent, manganese from 0.55 to 0.80 per cent, and phosphorus from 0.6 to 1.0 per cent.

Among seven irons in this range of composition, having from 0.50 to 0.67 per cent combined carbon, the coefficient of thermal expansion per degree centigrade from room temperature to 400°C. (750°F.) varied only from 0.0000125 to 0.0000128 and had no direct relation to phosphorus content. One iron, of 3.45 per

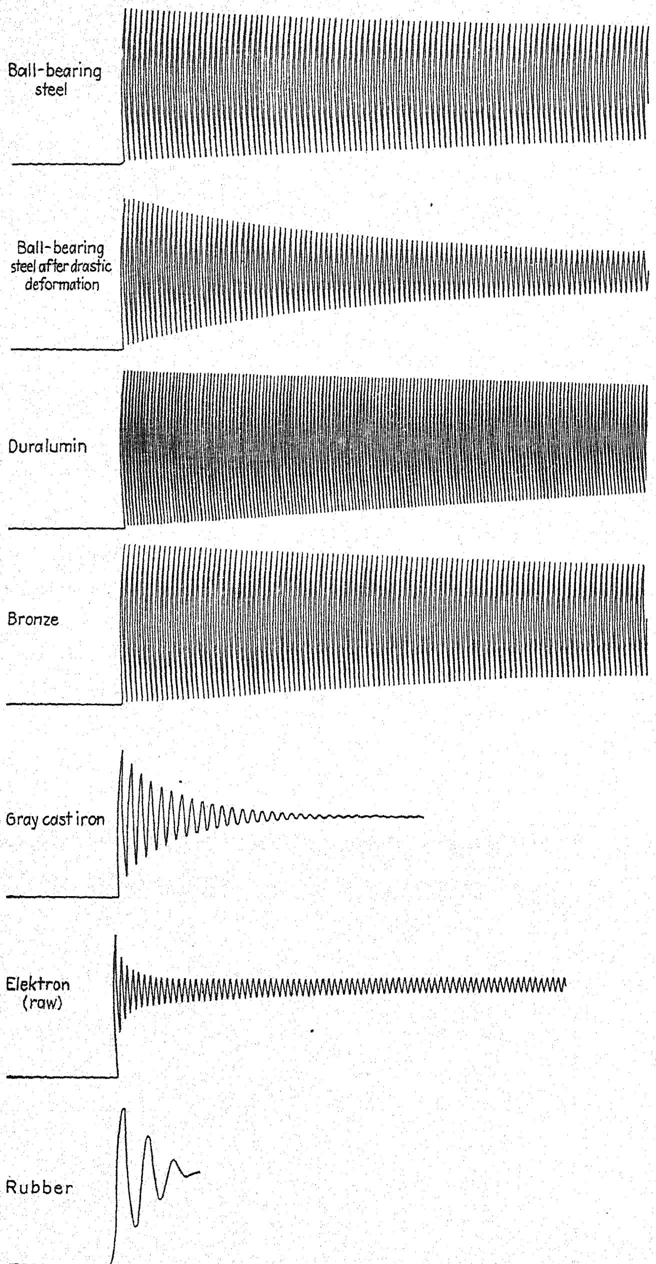

FIG. 89.—Damping curves of cast iron, alloy steel, non-ferrous alloys, and rubber. (*Knackstedt.*(383))

cent total carbon and only 0.35 per cent combined carbon with 0.59 per cent each of manganese and phosphorus, gave a coefficient of 0.0000123. It was therefore concluded that, to obtain a low coefficient, the combined carbon should be low.

The thermal conductivity of cast iron is given in the summary of the Symposium as ranging at room temperature from about 0.11 to 0.14 cal. per cm. per sec. per °C. It decreases as the temperature rises and is appreciably affected by changes in silicon content, the higher silicon irons having the lower conductivities. It is decreased slightly, other things being equal, as the combined carbon increases.

Söhnchen(775) has recently summarized the earlier literature on thermal and electric conductivities and added further experimental data. He gave higher values for thermal conductivity than those cited above, his range for gray iron being from 0.14 to 0.16 cal. per cm. per sec. per °C., that for black-heart malleable, 0.15 to 0.17, and for quenched cast iron, with its high combined carbon, 0.08 to 0.09. He stated that gray iron, cast steel, and cast red brass all have about the same thermal conductivity. Thermal conductivity of cast iron is further discussed in Chapter XV.

Electric conductivity varies directly with the silicon content and with the combined-carbon content and to about the same degree as does thermal conductivity. Söhnchen assembled his own data and those of some earlier workers in curves, from which the following approximate values are taken:

Total carbon, per cent	Silicon, per cent	Specific resistance, microhms per cu. cm.
2.5	1.0	49
3.0	1.0	51
3.5	1.0	70
2.5	2.0	70
3.0	2.0	80
3.5	2.0	100

Partridge's(276) data on electric conductivity were included in those summarized by Söhnchen and are also summarized in the Symposium on Cast Iron,(582) as are his data on magnetic proper-

ties. The magnetic properties of cast iron are briefly discussed in Chapter XVI.

D. AUTHOR'S SUMMARY

1. Cast iron is an impure iron-carbon-silicon alloy, usually containing 2 to 4 per cent carbon, 0.25 to 3 per cent silicon, and varying small amounts (usually less than 1 per cent) of manganese, sulphur, phosphorus, and other elements. Despite the fact that it is a ternary alloy, it is usually considered in industry—and is so considered in this and the next chapter—as an impure binary iron-carbon alloy with a variable small amount of silicon always present. Cast iron has been defined as an alloy of iron and carbon containing carbide eutectic, graphite eutectic, or both in the microstructure. For the purpose of this monograph the dividing line between cast iron and carbon steel has been placed at 1.70 per cent carbon.

2. The structure and properties of cast iron depend upon its composition, especially the carbon and silicon content, and upon the conditions under which it is melted and cast. Because of the important effect of melting and casting practices it is impossible to predict from the analysis of the raw materials used what the properties of a casting will be.

3. The mechanical and other properties of cast iron are closely related to its microstructure and to the appearance of the fracture. There are three primary constituents in cast iron: ferrite, cementite, and graphite. The appearance of the fracture—whether it is gray, mottled, or white—and the hardness of the material depend upon the relative amounts of these three constituents and upon the amount of pearlite. Steadite, a eutectic of iron phosphide, ferrite, and in some cases cementite, may also be present if the phosphorus is high enough. The ferrite in cast iron is not pure alpha iron; it is an iron-rich solution containing practically all of the silicon, and possibly a portion of some of the other elements, present. Cast iron may also contain non-metallic inclusions.

4. The hardness of unalloyed cast iron increases as the amount of cementite increases, but the strength does not necessarily increase with the hardness. The hardness depends chiefly upon the relative amounts of the constituents, but the strength and ductility depend not only upon the amount but also upon their

distribution, particularly upon the size and distribution of graphite. Gray cast iron may be considered as a low- or medium-carbon high-silicon steel in which graphite flakes are embedded. The form of these flakes greatly influences the ductility; hence, malleable iron, in which the graphite appears as rounded particles or nests, is much more ductile than ordinary gray iron, in which the graphite appears as flakes or plates.

5. Castings are usually made by melting cold pig iron and scrap cast iron, with or without the addition of steel scrap, in a cupola, air furnace, rotary furnace, or electric furnace, and pouring into loam, green-sand, dry-sand, or metal molds. Pig iron for foundry use is classified by the silicon content, which ranges from 1.25 to 3.25 per cent. Of the total pig-iron production in this country, about 12 per cent is used in the foundry and about 5 per cent is malleablized.

6. One of the most important characteristics of cast iron is its fluidity or running ability; in other words, its ability to fill a mold completely and thus reproduce accurately the design of the pattern. The fluidity is dependent chiefly upon the composition; it increases to a maximum with alloys of eutectic composition (4.3 per cent carbon with no silicon, or about 3.7 per cent carbon with 2 per cent silicon). According to some investigators, the fluidity also depends upon the amount of superheating, *i.e.*, the maximum temperature to which the molten iron was heated before pouring. High percentages of phosphorus lower the freezing point but, according to some authorities, this element does not increase fluidity if allowance is made for the lowering of the freezing temperature.

7. There is a marked decrease in volume as molten cast iron cools from a high temperature to the solidification point. At this latter temperature a slight increase in volume may result from the formation of graphite or the liberation of dissolved gas. Below the solidification point, the volume again decreases—but at a lower rate—as the iron cools to room temperature. The shrinkage decreases with the percentage of carbon from about 1.50 per cent for 2.25 per cent carbon to 0.75 per cent for 3.50 per cent carbon. The patternmaker's allowance for shrinkage varies from about $\frac{1}{10}$ or $\frac{1}{8}$ in. per ft. (0.75 or 1 per cent) for ordinary gray iron to $\frac{3}{16}$ or $\frac{1}{4}$ in. per ft. (1.5 or 2 per cent) for malleable iron.

8. The mechanical properties of gray iron cannot be altered to such a marked degree by alloying as is the case with steel since the discontinuities resulting from the graphite flakes limit the strength and ductility which can be attained. If pearlite is present, this constituent can be made harder by the introduction of carbide-forming elements, but these also increase the tendency of the iron to chill. Alloys which dissolve in the ferrite will, to a degree, strengthen and may toughen the iron.

9. White or chilled cast iron can be produced by controlling the carbon and silicon percentages and the cooling rate so that the fracture of the casting is all white or so that a white surface layer of the desired thickness is present. The tendency of molten cast iron of certain compositions to solidify with a white fracture when cooled rapidly is taken advantage of in the production of such materials as rolls where the surface layers must be white, hard, and wear resistant and the core soft, gray, and fairly ductile. Castings to be malleablized are poured from iron of such composition that they are white throughout. White and chilled irons cannot be readily classified on the basis of properties as the structure and properties vary from casting to casting and from one part to another of the same casting.

10. Ordinary gray cast iron is easily machinable unless it has a chilled surface layer or white or mottled areas in thin sections or at corners, or unless there is burned-in sand on the surface. According to the data available, there seems to be no definite relation between composition and structure of unchilled gray iron and its machinability as determined by lathe-tool tests. There seems to be, however, a direct relation between hardness and tensile strength and tool life as determined by the same tests. In one series of tests the machinability, as measured by tool life, decreased more than 50 per cent as the tensile strength increased from 17,000 to 50,000 lb. per sq. in.

11. The evidence regarding the effect of composition and other variables on the wear resistance of cast iron is conflicting. Boegehold has carefully considered all this evidence and offered the general conclusion that increased hardness, or the addition of elements which form hard constituents by combining chemically or which dissolve in and strengthen the ferrite, is accompanied by decreased wear. Boegehold recommended that, in the future,

work on wear testing be directed toward finding causes for the exceptions to these general conclusions.

12. When cast iron is repeatedly heated and cooled, or when it is held at a high temperature, there is a permanent increase in volume. This growth, which may amount to as much as 50 per cent, seems to be caused primarily by graphitization and by oxidation along the graphite flakes. These result in the formation of fine cracks and a mechanical swelling. It seems likely that stresses introduced by thermal changes or differences in the expansion of iron, graphite, and non-metallic inclusions with changes in temperature contribute to the formation of the cracks and the increase in volume. The tendency of cast iron to grow increases directly with the silicon content or the content of other graphitizing elements. Elements which retard graphitization (for example, chromium) retard growth. Oberhoffer and Piwowarsky found practically no tendency for growth in a silicon-free alloy containing 4.3 per cent carbon when it was repeatedly heated to temperatures which caused growth in normal cast iron.

13. The possibility of hot working cast iron slightly while it is heated into the austenitic range, the difficulties in welding this material, and its superior ability to damp out vibration are discussed briefly. A few data on the miscellaneous physical properties, including thermal expansion, thermal conductivity, and electric conductivity, are also given. The coefficient of thermal expansion increases as the carbon decreases. Thermal conductivity decreases with increasing temperature and is appreciably affected by the amount of silicon present. Electric resistivity varies directly with the amount of silicon and combined carbon. Other data on the physical properties of cast iron are included in the discussion of the physical properties of iron-carbon alloys in Chapters XV and XVI.

CHAPTER IX

THE MECHANICAL PROPERTIES OF CAST IRON

Strength Properties—Relation between Composition, Structure, and Mechanical Properties—Effect of Superheating and Ladle Additions—Effect of Other Variables on Mechanical Properties—Heat Treatment of Gray Cast Iron—Author's Summary

Until recently gray cast iron, as compared with ordinary structural steel, was a brittle material of relatively low tensile strength. As a material of construction it has the advantage of cheapness and of being readily cast into any desired shape; it is exceedingly valuable in some engineering applications because of its high compressive strength, its insensitivity to the effect of notches especially under repeated stress, and, as noted in the previous chapter, its ability to damp out vibration quickly. The disadvantage of low tensile strength is no longer serious; with proper attention to composition and foundry technique it is now possible to produce gray-iron castings whose tensile strength approaches that of low-carbon steel. Because of the graphite always present, even in high-strength irons, tensile specimens show little or no permanent elongation; hence, the relative brittleness of cast iron must be given full consideration by designers.

There are many factors which influence the mechanical properties of cast iron. In addition to the chemical composition which is the most important, the type and temperature of the mold, the size of the section cast, the maximum temperature of the metal prior to pouring, and the temperature of the metal when cast all affect the properties. Thus it is evident that, although differences in composition have the most pronounced effect, gray cast irons of practically the same analysis may vary widely in mechanical properties.

Except as material for the production of malleable castings, white cast iron is of little industrial importance. Moreover, white cast irons cannot be classified readily on the basis of

mechanical properties; hence, there are relatively few data available. These few are summarized on pages 334 and 335 of this chapter. Nearly all of the other data correlated in this chapter were obtained on gray iron.

A. STRENGTH PROPERTIES

There are five principal tests made to determine the strength properties of gray cast iron. These are tests for: (*a*) transverse strength, (*b*) tensile strength, (*c*) compressive strength, (*d*) endurance limit, and (*e*) impact resistance. In the following sections the importance of these tests is briefly discussed and the effects of composition and other variables on the properties of cast iron, as shown by these tests, are given in some detail.

105. Transverse Strength.—It is the usual procedure to judge the quality of gray cast iron by the results of tensile and transverse tests. The latter are ordinarily made on unmachined round bars. The size of the bars depends upon the controlling section of the casting and, according to a tentative specification (A48-32T) issued in 1932 by the American Society for Testing Materials, is as follows:

Controlling section of casting, in.	Normal diameter of test bar as cast, in.	Length of test bar, in.	Distance between supports, in.
0.75 max.	0.875	15	12
0.76 to 1.10	1.20	21	18
1.11 to 2.00	2.00	27	24

An earlier specification called for bars 1.25 in. in diameter, tested on supports 12 in. apart. In making transverse tests, it is customary to determine the deflection of the center of the bar; the values of deflection, usually for deflection at fracture, are used as an index of flexibility.

In this country, the transverse strength is ordinarily reported as the load required to fracture a specimen of a given diameter broken on supports a certain distance apart. If the diameter of the specimen varies slightly from the desired diameter, the observed value is corrected to give the load which would be required to break a bar of exactly the desired diameter. Another method of reporting transverse strength, which is used abroad,

is to report what is called "modulus of rupture." This, according to the Symposium on Cast Iron,[582] is "a mathematical convention adopted for expressing the transverse breaking strength (flexural strength) of materials which are of low ductility and which, in general, depart from Hooke's law, which is, that 'stress divided by strain is constant.'"

The general formula for modulus of rupture (MR) for a bar loaded at the center is

$$MR = \frac{Plc}{4I}$$

where P is the breaking load, l the distance between supports, c the distance from the neutral axis to the extreme fiber, and I the moment of inertia. If the load is in pounds and the distances are in inches, the modulus of rupture will be expressed as pounds per square inch. When the metric system is used, the modulus is reported in kilograms per square millimeter.

Heller and Jungbluth[715] pointed out that the conversion by formula of results obtained with the 12- or 18-in. span or those obtained on the German standard bar [30-mm. (1.2-in.) diameter and 600-mm. (23.6-in.) span] into results for longer or shorter spans is not entirely accurate when experimentally checked. In the case of bars of different diameters the size effect, *i.e.*, a difference in actual structure and properties, also enters. Heller and Jungbluth gave correction factors for some grades of iron.

In general, the transverse strength of a machined bar will be slightly different from that of an unmachined bar,[277] and the difference in strength will depend on the amount of metal removed, on the character of the surface, and on the chilling tendency of the iron. Results later obtained by Meyersberg,[552] however, indicated that there was very little difference in the modulus of rupture between bars cast to 30 mm. (1.2 in.) diameter and those cast to 34 mm. (1.34 in.) and turned to 30 mm., *i.e.*, when only a little metal is removed. Meyersberg tested 94 unmachined bars and 97 machined bars and showed his results as frequency curves. There was no important difference in the frequency curves for the two sets of specimens. The effects of size of section and of the casting "skin" are discussed on page 328.

MacKenzie[737] stated that the transverse test is the most readily accessible source of information on strength, flexibility, toughness, damping capacity, plastic deformation, surface condition, and machinability. It is becoming more customary recently to machine tensile specimens from the broken ends of the transverse-test bar and secure tensile data as well.

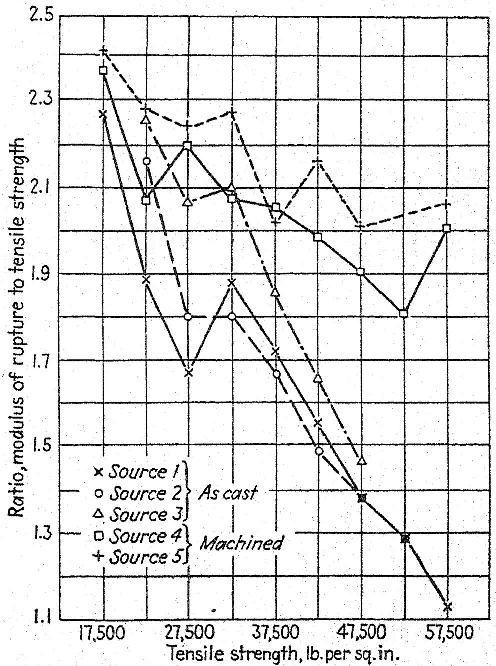

FIG. 90.—Relation of tensile strength to ratio of modulus of rupture to tensile strength. Sources of the data are: (1) Symposium Committee's correlation of data on plain cast iron as reported in American literature, 1923 to 1933; (2) MacKenzie[470]; (3) Thum and Ude[348]; (4) Subcommittee XV of Committee A-3, A.S.T.M.,[586] from 1.20-in. transverse and 0.8-in. tensile specimens, machined; (5) same as (4), from 0.5 × 5-in. transverse and 0.5-in. tensile specimens, machined. (*Symposium on Cast Iron.*[582])

106. Tensile Strength and Modulus of Elasticity.—The tentative specification of the American Society for Testing Materials mentioned on page 289 lists seven grades of gray iron, classified by tensile strength. In this list, the class number of the iron, 20, 25, 30, 35, 40, 50, and 60, is the minimum tensile strength in thousand lb. per sq. in. This specification also calls attention to the fact that there is no simple relationship between tensile and

transverse strengths because of the deviation from Hooke's law. As pointed out in the Symposium on Cast Iron,[582] the factors which cause deviations from Hooke's law also tend to decrease strength, and the lower the strength the greater is the difference between tensile strength and modulus of rupture. Figure 90, from the Symposium, shows the relation of tensile strength to the ratio of modulus of rupture to tensile strength for irons of different

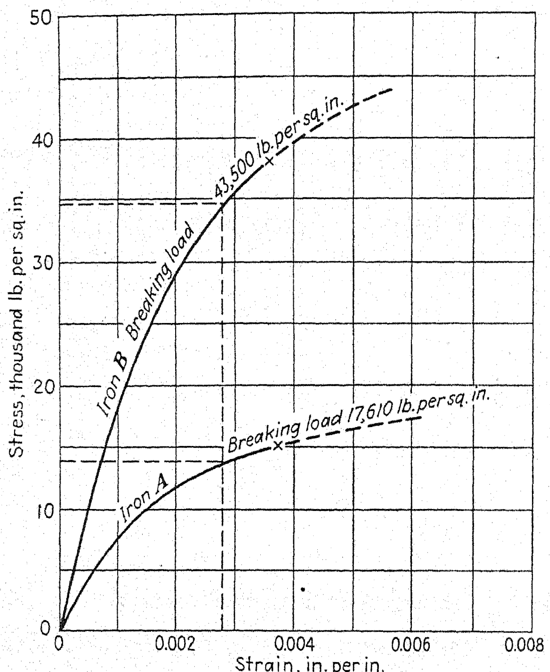

FIG. 91.—Stress-strain curves of two cast irons tested in tension. (*Bolton.*[510])

strengths. It shows that as the tensile strength increases the ratio decreases, but that there is considerable spread in the values.

Stress-strain curves for soft cast irons are generally curved from the origin as shown in Fig. 91 from Bolton.[510] Iron *B* was a high-strength iron and iron *A* a low-strength iron produced by pouring an iron suitable for light castings into a heavy section. As strain is not proportional to stress even in the range of low stresses, cast iron cannot be said to have a true modulus of elasticity. (Further information on elastic constants is given in Chapter XV.) What is usually termed modulus of elasticity

means the relative stiffness under particular conditions of loading. The effective modulus of elasticity at 25 per cent of the tensile strength ranges from 12,000,000 lb. per sq. in. for the weaker irons to 18,000,000 lb. per sq. in. or more for the stronger type of gray irons.(582)

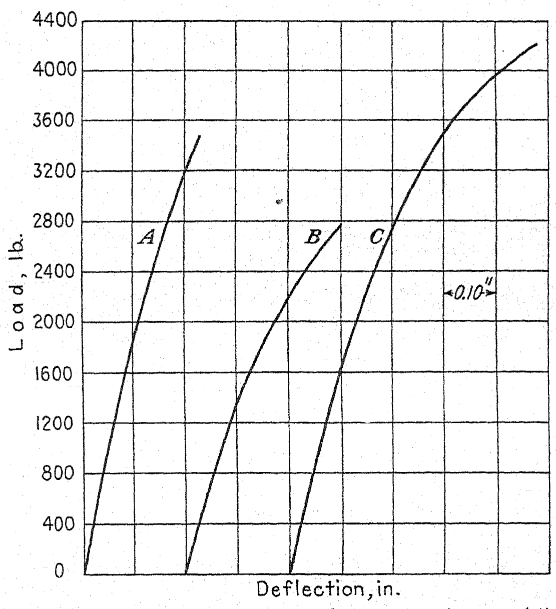

FIG. 92.—Transverse stress-strain curves of three cast irons. (A) = "Stiff" cast iron, 1 per cent combined carbon, 1.5 per cent graphite; (B) = "Ordinary" cast iron, 0.68 per cent combined carbon, 2.44 per cent graphite; (C) = 0.85 per cent combined carbon, 1.95 per cent graphite. (*MacKenzie.*(737))

Wickenden and Vanick, in commenting upon this portion of the text,* called attention to an exception to the generally accepted assumption that gray cast iron has no definite modulus of elasticity:

The statement that stress-strain curves for cast iron are curved from the origins is not always true. By relieving the stress in specimens at about 540°C. (1000°F.) the stress-strain curve of some high-strength irons becomes straight up to 12,000 or 14,000 lb. per sq. in. Ordinary soft irons show this with less certainty, but the curve may be straight up to about 4000 lb. per sq. in. On high-strength iron of 60,000 lb. per sq. in. tensile strength, the modulus of elasticity may be between 22 and 24 million.

* Private communication.

Curves similar in general characteristics to those shown in Fig. 91 are given in Fig. 92, from MacKenzie.(737) These are transverse stress-strain curves for three types of gray iron, and all curve from the origins. Figure 93, from Thum,(347) shows

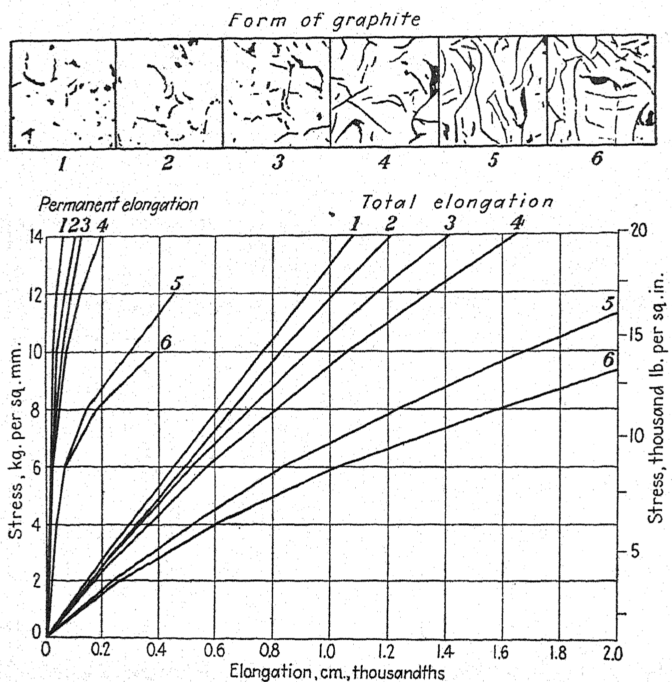

FIG. 93.—Relation between stress-strain curves and size of graphite flakes. (*Thum.*(347))

stress-strain curves for several gray cast irons and the distribution of graphite in the materials. The effective modulus of elasticity of these irons was given by Thum as follows:

Iron number	Modulus of elasticity, kg. per sq. cm.	Modulus of elasticity, Million lb. per sq. in.
1	1,380,000	19.6
2	1,250,000	17.8
3	1,160,000	16.5
4	1,145,000	16.3
5	895,000	12.7
6	759,000	10.8

As the size of the graphite flakes increases, the stress-strain curves approach the strain axis, which is equivalent to a decrease in effective modulus of elasticity.

The amount of graphite as well as the size of the particles affects the modulus; as the content and size of the particles decrease the modulus approaches more nearly that of steel. Malleable iron has a modulus of about 25 million lb. per sq. in., and in some of the heat-treated cast irons or "quick-annealed" malleables the moduli may be as high as 28 million. White cast iron which contains no graphite may have a modulus as high as 28 million lb. per sq. in.

107. Compressive Strength.—The compressive strength of cast iron is several times the tensile strength and, as shown in Table 65 from the Symposium on Cast Iron(582) (which summarizes the work of an A.S.T.M. Committee(586)), the ratio of compressive strength to tensile strength decreases as the latter increases.

TABLE 65.—RELATION BETWEEN COMPRESSIVE AND TENSILE STRENGTHS*

Tensile strength, lb. per sq. in.	Number of tests	Average ratio
15,000 to 20,000	4	4.08
20,100 to 25,000	4	4.02
25,100 to 30,000	2	3.68
30,100 to 35,000	7	3.61
35,100 to 40,000	3	3.39
40,100 to 45,000	2	2.99
45,100 to 50,000	1	3.31†
50,100 to 55,000	1	3.08†
55,100 to 60,000	1	2.45†

* Symposium on Cast Iron.(582)

† Vanick, in a private communication, suggested that a better average would be 2.95 for the whole group of 45,000 to 60,000 lb. per sq. in. tensile strength.

For many industrial applications, such as bases for machinery, pipes used as structural members, and many others, cast iron is used for its strength in compression. Some of these parts are so massive in order to supply rigidity that the strength of even the weakest cast iron is ample. In machine bases the damping ability (see page 281) of cast iron may also be of value. This property is associated with the discontinuities formed by the graphite flakes and with the resultant inelastic behavior, which

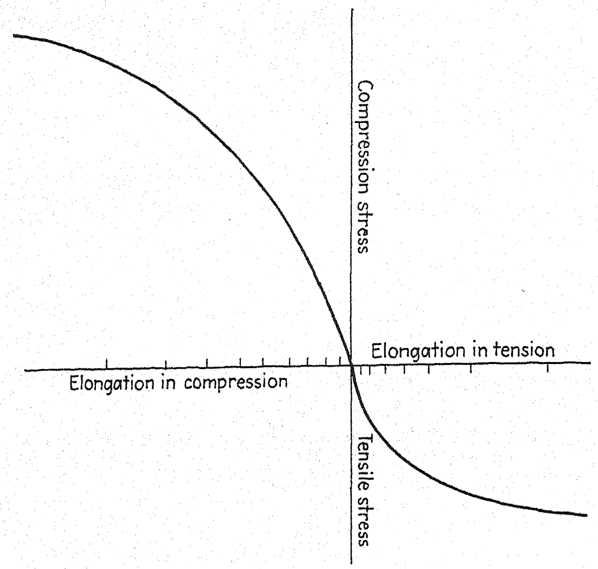

FIG. 94.—Stress-strain curve in tension and compression of cast iron of 17,000 lb. per sq. in. tensile strength. (*Bach and Baumann.*[112])

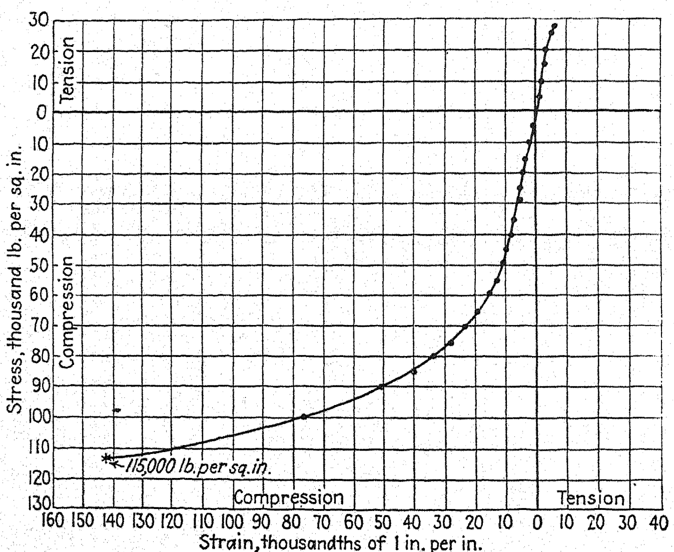

FIG. 95.—Stress-strain curve of gray cast iron in tension and compression. (*Bolton.*)

is indicated by the shape of the stress-strain curve (see also Chapter XVII).

Stress-strain curves in both tension and compression are shown in Fig. 94 from early work of Bach(112) on an iron of 17,000 lb. per sq. in. tensile strength. Curves of the same type are plotted in Figs. 95 and 96 by Bolton* who used a low-strength iron of the following composition:

Element	Percentage
Total carbon	3.37
Combined carbon	0.34
Silicon	2.56
Manganese	0.65
Sulphur	0.057
Phosphorus	0.30
Copper	0.14

For compressive strengths up to 100,000 lb. per sq. in. (and for tensile strength) Bolton used an 0.8-in. diameter bar, 3 in. long, machined from a 1.2-in. bar cast in sand. For higher compressive loads a specimen 0.8×1 in. was used. The mechanical properties of the iron were as follows:

Tensile strength, lb. per sq. in.	26,800
Compressive strength, lb. per sq. in.	115,000
Transverse modulus of rupture, lb. per sq. in.	61,200
Deflection on 18-in. span, in.	0.275
Shear, on 0.505-in. bar, lb. per sq. in.	40,925
Modulus of rupture in torsion, 1.14-in. bar, lb. per sq. in.	39,000
Brinell hardness	185

Figure 95 shows the tension-compression stress-strain curves and Fig. 96 a section of both curves, for the low stresses, on a larger scale. The engineering significance of the curved stress-strain line for gray cast iron has been discussed recently by Hurst(722) and by Lessells.(733)

108. Endurance Limit and Impact Resistance.—As mentioned on page 288, one of the marked advantages of cast iron as a structural material is its relative insensitivity to the effect of notches and other surface imperfections when subjected to repeated stress. This notch insensitivity and the endurance limit, together with a brief survey of endurance and damping of

* Private communication.

cast iron, are included in the discussion of the behavior of iron-carbon alloys under repeated stress (Chapter XI, section 151).

Cast iron has a low resistance to single-blow and repeated impact; moreover, with the usual tests the results are not trustworthy; hence, the only discussion of the impact resistance of

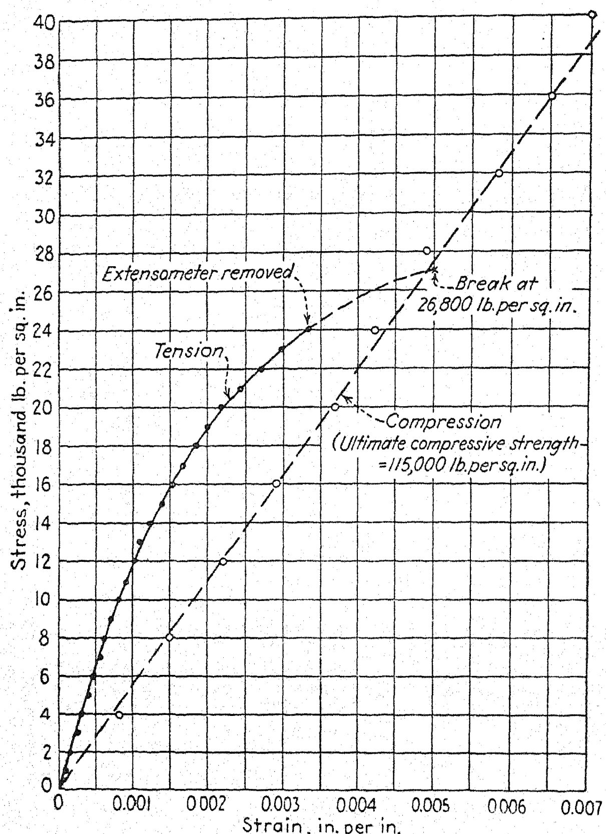

FIG. 96.—Same as Fig. 95 near zero stress and strain ordinates. (*Bolton.*)

cast iron which can be given in this section is reproduction of the conclusions drawn from a complete study of impact testing of cast iron as reported by a Committee of the American Society for Testing Materials.(586) These are as follows:

1. The very small specimens, such as are used for steel, are not reliable indices of the impact strength of cast iron. Enger's modified Izod with

a fairly large specimen is by far the best used, though there is no reason to believe the same size specimen would not give equally good, and perhaps better, results in a Charpy machine.

2. The Krupp-Stanton test gave relatively enormous differences in results for irons closely approaching each other in all other properties. It is difficult to see how this test means anything when applied to cast iron.
3. All notched specimens were unsatisfactory.
4. The tensile impact test does not seem to be a reliable index of impact value.
5. When using specimens 1.20 in. in diameter, the large pendulum machines of the Russell and Charpy type give results which follow closely the resilience measured in the transverse test.
6. The drop tests seem to indicate the impact resistance left in the bar after previous loading to a point just below the breaking point. A drop test, therefore, considered in the light of the transverse resilience would appear to be a useful index of the relative amounts of plastic and elastic deflection.
7. The bending curve of the transverse test, if the test is accurately carried out, is capable of giving most of the information needed about cast iron, and the Brinell hardness determined over the whole cross-section is the most valuable supplementary test.
8. The drop-test machines differ greatly in their construction. They would probably become quite generally used if the machines were carefully standardized in all dimensions. The tests would then give valuable corroboration to the results of the transverse tests.
9. The committee does not recommend at this time any move toward an impact requirement in specifications.

A discussion of the variables which affect the impact resistance, particularly of carbon steels, is given in Chapter XVII.

B. RELATION BETWEEN COMPOSITION, STRUCTURE, AND MECHANICAL PROPERTIES

As has been repeatedly stated in this volume and in Volume I of this monograph, gray cast iron may be considered as a steel matrix interspersed with numerous particles of graphite. The properties of the iron are, therefore, dependent upon the characteristics of the "steel" matrix and upon the size and distribution of the graphite. It has been emphasized by a number of investigators in the past, and especially by Bardenheuer and Zeyen,[(248)] that the size and distribution of the graphite have more influence on the strength of iron than have the properties of the "steel"

matrix. To obtain high strength the iron must not contain large graphite flakes; on the other hand, a high-strength iron may contain almost no combined carbon provided the graphite is finely disseminated. With highly dispersed graphite the strength of an iron is increased by increasing the amount of combined carbon up to the content required to form a matrix consisting entirely of pearlite, but producing a pearlitic matrix in an iron containing large flakes of graphite increases the strength but slightly, even though the hardness is higher. The problem of producing high-strength gray iron, therefore, resolves itself

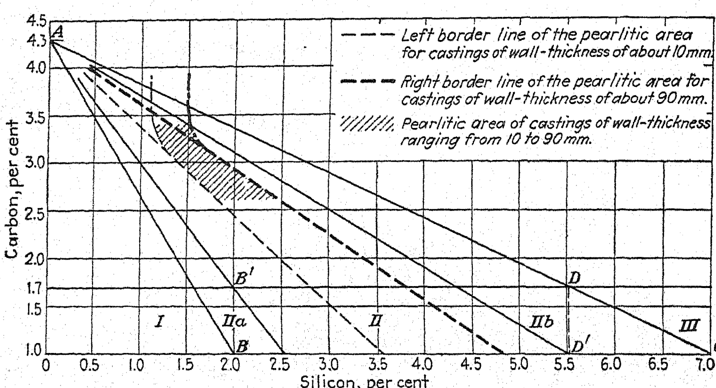

FIG. 97.—Composition limits for various structures in cast iron. (*Maurer and Holtzhaussen.*(219))

into the problem of producing highly dispersed graphite in a cast iron with a pearlitic matrix.

109. Influence of Composition on Structure—The Maurer Diagram.—The structure of gray-iron castings is controlled chiefly by regulating the carbon and silicon contents of the iron. In order to produce gray iron, these elements must be held within certain limits in the melt; these limits may, for a definite cooling rate, be roughly indicated by a structural diagram such as was originally proposed by Maurer.*(130) The Maurer diagram, shown in Fig. 97, resembles the well-known structural diagrams constructed by Guillet to show the structures found in alloy steels.

Disregarding the dashed lines and shaded area, which do not form a part of the original diagram, Fig. 97 shows three primary

* The Maurer and the Maurer-Holtzhaussen diagrams are discussed in a former monograph of this series.(613)

and two secondary fields; the secondary fields are for alloys showing structures of the adjoining primary fields. Alloys to the left of line AB, field I, solidify as white irons. Alloys in field II, between lines AB' and AD', form gray iron containing no free ferrite, this being the field of pearlitic irons. Alloys to the right of line AC, field III, form gray iron containing no pearlite. Of the secondary fields, II*a* is the field in which mottled iron is formed, and II*b* is the field in which iron containing both pearlite and free ferrite is formed.

The diagram as described above is for castings of test-bar size sand cast at normal casting temperatures from commercial irons. As the size of the section increases, the lines of the diagram shift to the left, and for castings in chill molds the lines shift toward the right. The dashed lines and the shaded area were drawn by Maurer and Holtzhaussen[219] to show a field for alloys which would be pearlitic when cast in sections having a thickness of not less than 10 mm. (0.4 in.) and not more than 90 mm. (3.5 in.).

As a result of several investigations, since the Maurer-Holtzhaussen diagram (Fig. 97) was published in 1927, the various structural fields shown in Fig. 97 have been changed. In one of the most recent of these investigations, by Uhlitzsch and Weichelt,[667] a marked change in the ferritic field is indicated. According to these investigators, the line AC (Fig. 97) should be placed at higher silicon contents. For example, for small sections cast in sand the line should pass through the 5.5 per cent silicon ordinate just below the 3 per cent carbon line. Accordingly a wholly ferritic structure is possible only with higher silicon percentages than are indicated in Fig. 97. Uhlitzsch and Weichelt also investigated the effect of the wall thickness on the location of the pearlitic field.

Maurer and Holtzhaussen themselves showed that their diagram (Fig. 97) was not entirely accurate in outlining the fields for normal irons cast as test bars 30 mm. (1.2 in.) in diameter. However, as the diagram should be regarded only as a rough approximation, it was not considered advisable to change the original diagram. Maurer and Holtzhaussen prepared about a hundred melts in which the carbon content ranged from 2.4 to 3.8 per cent and the silicon content from 0.8 to 5 per cent. All of the irons contained from 0.8 to 1 per cent manganese, 0.3 per cent phosphorus, and 0.1 per cent sulphur. Test bars 30 mm.

(1.2 in.) in diameter were cast into dry-sand molds at normal temperatures, at 250°C. (480°F.), and at 450°C. (840°F.). Samples were also cast into chill molds. The tensile properties of the samples cast in sand molds at different temperatures are shown in Fig. 98. Each plotting point shows the strength of one specimen. In general, alloys of high strength were in the pearlitic field II. Brinell-hardness values of the specimens cast

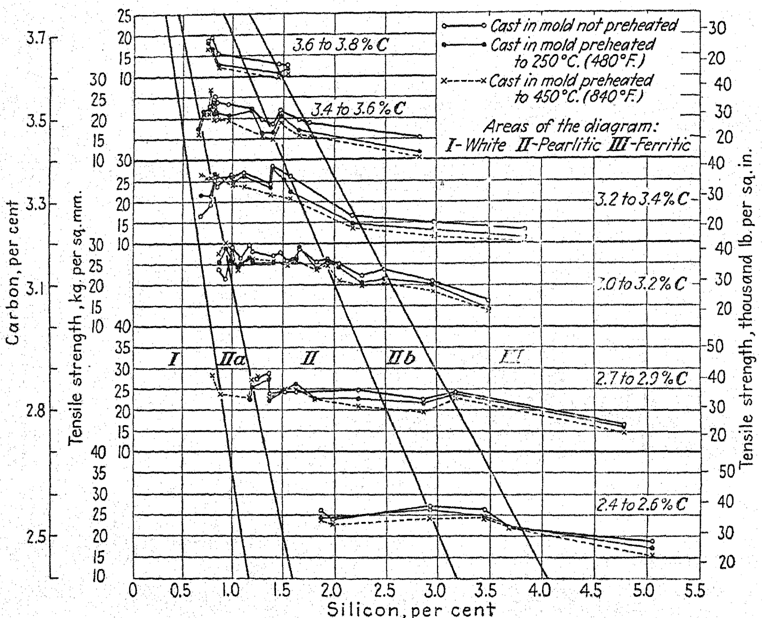

FIG. 98.—Relation between tensile strength and structure of iron cast into sand molds at different temperatures. (*Maurer and Holtzhaussen.*[219])

into unheated molds are shown by two views of a space model, Figs. 99 and 100, and by the curves in Fig. 101. The height of the model is proportional to the hardness; the lines upon it show the fields outlined by Maurer's diagram. The double circles in Fig. 101 indicate intersections of the constant-carbon sections with lines on the diagram. The abrupt increase in hardness when the field of white iron is entered is clearly shown by the space model.

110. Other Data on the Influence of Composition on Properties. Some results recently reported by Koch and Piwowarsky[628]

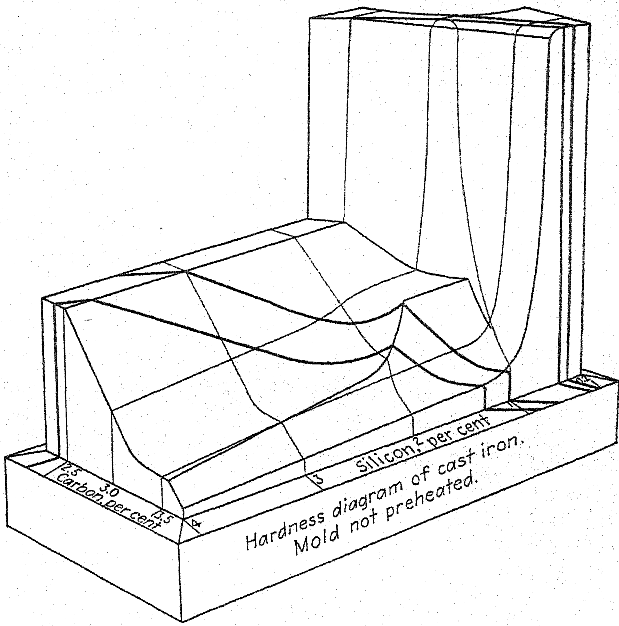

FIG. 99.—Space model showing relation between carbon and silicon percentages and Brinell hardness of cast-iron test bars. (*Maurer and Holtzhaussen.*[219])

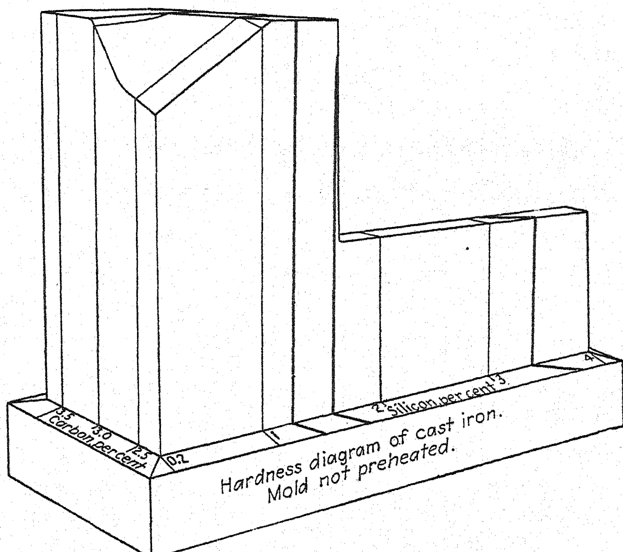

FIG. 100.—Another view of model shown in Fig. 99. (*Maurer and Holtzhaussen.*[219])

indicate how the mechanical properties of gray iron vary with the carbon and silicon contents and with the size of section into which it is cast. The tests were made on the 12 irons whose

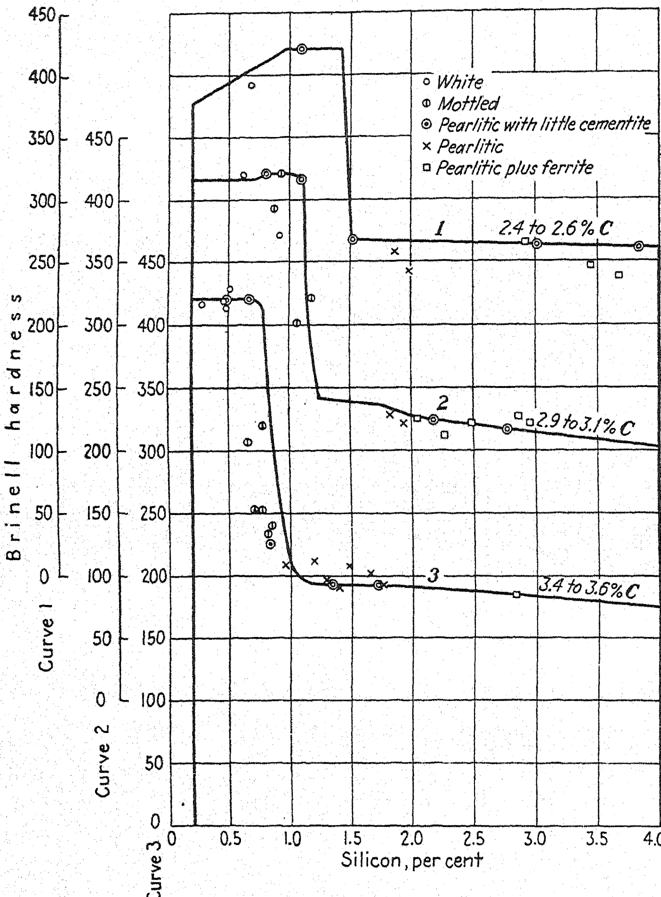

FIG. 101.—Influence of silicon on the Brinell hardness of cast irons containing: (1) 2.4 to 2.6 per cent carbon, (2) 2.9 to 3.1 per cent carbon, and (3) 3.4 to 3.6 per cent carbon. (*Maurer and Holtzhaussen.*[219])

compositions are shown in Table 66. These irons were made in a Brackelsberg furnace (a rotating coal-fired type) from mixtures of a special German pig iron of high purity, a hematite iron, scrap rails, and ferrosilicon. As will be observed from the analyses, the sulphur, phosphorus, and manganese contents

were substantially constant. The irons had comparatively low strengths. From iron of each composition samples were cast at temperatures of 1440 and 1340°C. (2625 and 2445°F.). Four test bars 650 mm. (26 in.) long and of different diameters were cast into the same dry-sand mold, the bars being cast

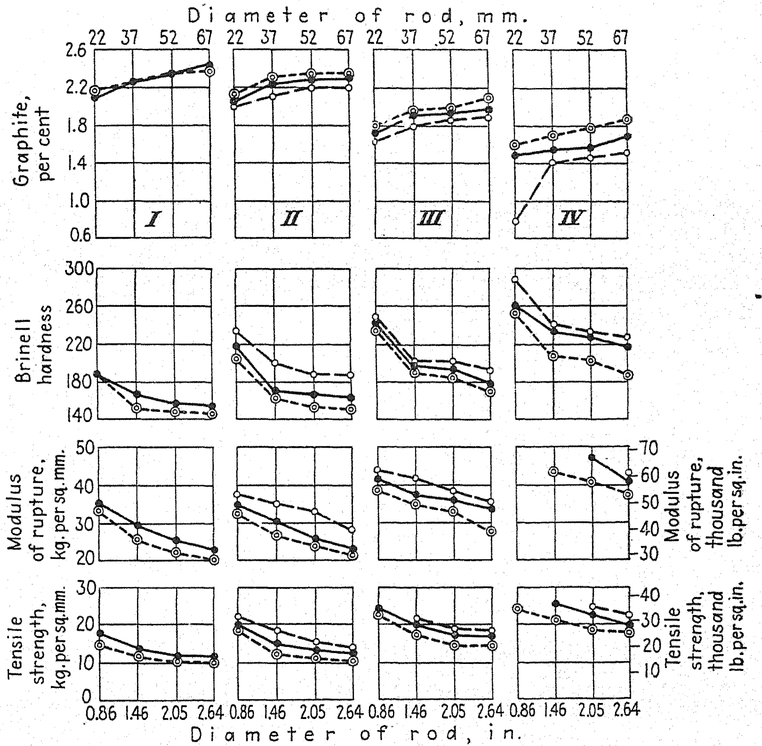

Fig. 102.—Properties of cast irons of different compositons (see Table 66), cast into bars of different diameters. (*Koch and Piwowarsky.*[628])

vertically and fed from the bottom. The cast bars had diameters of 22, 37, 52, and 67 mm. (0.86, 1.46, 2.05, and 2.64 in.). For the transverse test the bar of the smallest diameter was not machined, while the other bars were turned to a diameter of 30 mm. (1.2 in.). The bars were broken on a 600-mm. (23.6-in.) span. The results of tensile, transverse, and hardness tests on all samples are given in Table 67. The results of some of the tests made on samples cast at 1440°C. (2625°F.) are plotted in Fig. 102 which shows that strength and hardness increase as

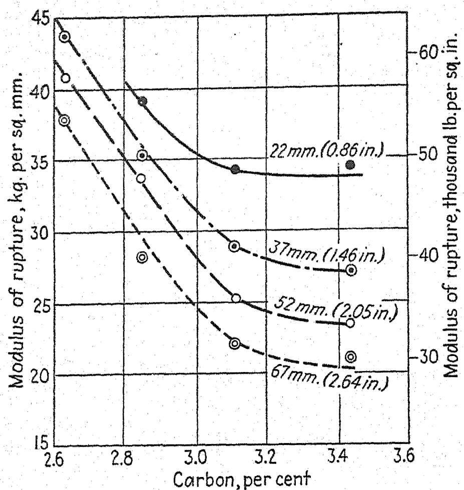

FIG. 103.—Effect of carbon content on the modulus of rupture of cast-iron bars of various sizes containing 2.25 per cent silicon. (*Koch and Piwowarsky.*[628])

TABLE 66.—COMPOSITION OF IRONS MADE BY KOCH AND PIWOWARSKY[628]

Group	Identification in Fig. 102	Composition, per cent				
		Total carbon	Si	Mn	P	S
I	. . . .	3.51	1.64	0.40	0.060	0.042
	Solid line	3.44	2.03	0.38	0.061	0.043
	Short dashes	3.40	2.51	0.39	0.064	0.040
II	Long dashes	3.12	1.65	0.40	0.069	0.043
	Solid line	3.12	2.00	0.41	0.076	0.042
	Short dashes	3.03	2.52	0.38	0.072	0.041
III	Long dashes	2.82	1.63	0.47	0.066	0.041
	Solid line	2.84	1.85	0.47	0.066	0.040
	Short dashes	2.85	2.23	0.48	0.068	0.044
IV	Long dashes	2.59	1.30	0.43	0.069	0.045
	Solid line	2.61	1.69	0.45	0.068	0.046
	Short dashes	2.63	2.24	0.41	0.073	0.042

TABLE 67.—RELATION BETWEEN PROPERTIES OF CAST IRON, COMPOSITION, CASTING TEMPERATURE, AND SIZE OF TEST BAR*

Number	Diameter, mm.	Diameter, in.	Composition, per cent: Total carbon	Si	Graphite	Combined carbon	Tensile strength, lb. per sq. in.	Transverse test†: Modulus of rupture, lb. per sq. in.	Deflection, in.	Brinell hardness
Group I. Cast at 1440°C. (2625°F.)										
1A	22	0.86	3.51	1.64	2.23	1.28	24,200	43,700	0.61	187
1B	37	1.46			2.26	1.25	17,200	42,000	0.51	179
1C	52	2.05			2.57	0.94	14,000	33,000	0.45	143
1D	67	2.64			2.72	0.79	11,500	25,100	0.41	121
2A	22	0.86	3.44	2.03	2.10	1.34	24,600	50,000	0.67	187
2B	37	1.46			2.25	1.19	19,200	41,600	0.49	165
2C	52	2.05			2.36	1.08	16,100	35,600	0.45	156
2B	67	2.64			2.43	1.01	15,100	31,900	0.47	153
3A	22	0.86	3.40	2.51	2.16	1.24	20,900	48,000	0.59	187
3B	37	1.46			2.25	1.15	16,800	36,200	0.51	152
3C	52	2.05			2.36	1.04	15,100	31,300	0.49	149
3D	67	2.64			2.40	1.00	14,000	27,800	0.43	146
Group I. Cast at 1340°C. (2445°F.)										
1AK	22	0.86	3.51	1.64	2.33	1.18	26,600	43,800	0.53	187
1BK	37	1.46			2.37	1.14	20,400	41,300	0.43	170
1CK	52	2.05			2.38	1.13	15,800	37,200	0.47	146
1DK	67	2.64			2.40	1.11	12,700	27,500	0.43	131
2AK	22	0.86	3.44	2.03	2.16	1.28	25,200	41,400	0.55	187
2BK	37	1.46			2.20	1.24	19,200	42,000	0.53	161
2CK	52	2.05			2.37	1.07	17,100	36,600	0.49	156
2DK	67	2.64			2.39	1.05	15,900	33,700	0.49	153
3AK	22	0.86	3.40	2.51	2.12	1.28	20,400	34,500	0.47	187
3BK	37	1.46			2.18	1.22	18,400	39,200	0.57	157
3CK	52	2.05			2.32	1.08	15,500	33,900	0.49	149
3DK	67	2.64			2.37	1.03	14,500	29,900	0.51	143
Group II. Cast at 1440°C. (2625°F.)										
4A	22	0.86	3.12	1.65	2.01	1.11	31,000	53,100	0.35	235
4B	37	1.46			2.12	1.00	26,000	49,800	....	202
4C	52	2.05			2.20	0.92	21,800	46,800	0.45	187
4D	67	2.64			2.22	0.90	19,100	35,900	0.37	183
5A	22	0.86	3.12	2.00	2.07	1.05	28,800	50,100	0.43	217
5B	37	1.46			2.24	0.88	21,400	44,000	0.47	170
5C	52	2.05			2.28	0.84	18,900	36,300	0.39	166
5D	67	2.64			2.30	0.82	17,800	32,600	0.37	163
6A	22	0.86	3.02	2.52	2.12	0.91	28,200	46,400	0.41	207
6B	37	1.46			2.28	0.75	18,100	38,200	0.41	163
6C	52	2.05			2.31	0.72	16,500	34,700	0.39	156
6D	67	2.64			2.32	0.71	15,500	30,200	0.37	153

TABLE 67.—RELATION BETWEEN PROPERTIES OF CAST IRON, COMPOSITION, CASTING TEMPERATURE, AND SIZE OF TEST BAR.*—(*Continued*)

Number	Diameter		Composition, per cent				Tensile strength, lb. per sq. in.	Transverse test†		Brinell hardness
	mm.	in.	Total carbon	Si	Graphite	Combined carbon		Modulus of rupture, lb. per sq. in.	Deflection, in.	
Group II. Cast at 1340°C. (2445°F.)										
4*AK*	22	0.86	3.12	1.65	1.98	1.14	29,200	52,500	0.39	241
4*BK*	37	1.46			2.10	1.02	26,500	57,100	0.53	207
4*CK*	52	2.05			2.15	0.97	22,900	52,400	0.49	179
4*DK*	67	2.64			2.20	0.92	22,100	45,300	0.47	174
5*AK*	22	0.86	3.12	2.00	2.02	1.10	31,200	52,800	0.39	235
5*BK*	37	1.46			2.22	0.90	21,800	51,000	0.47	183
5*CK*	52	2.05			2.26	0.86	19,500	42,300	0.43	170
5*DK*	67	2.64			2.28	0.84	18,100	37,500	0.43	170
6*AK*	22	0.86	3.03	2.52	2.05	0.98	29,300	54,800	0.47	228
6*BK*	37	1.46			2.03	1.00	24,900	53,800	0.49	202
6*CK*	52	2.05			2.04	0.99	19,900	46,300	0.47	183
6*DK*	67	2.64			2.09	0.94	19,800	42,400	0.45	183
Group III. Cast at 1440°C. (2625°F.)										
7*A*	22	0.86	2.82	1.63	1.66	1.16	32,600	62,200	0.47	248
7*B*	37	1.46			1.80	1.02	31,000	59,100	0.45	207
7*C*	52	2.05			1.86	0.96	27,000	54,500	0.45	202
7*D*	67	2.64			1.89	0.93	25,500	50,200	0.43	192
8*A*	22	0.86	2.84	1.86	1.73	1.11	34,300	59,700	0.45	241
8*B*	37	1.46			1.92	0.92	29,600	53,500	0.45	196
8*C*	52	2.05			1.94	0.90	25,200	51,800	0.43	192
8*D*	67	2.64			1.97	0.87	24,300	47,200	0.43	179
9*A*	22	0.86	2.85	2.23	1.80	1.05	32,600	55,500	0.33	240
9*B*	37	1.46			1.96	0.89	24,900	49,900	0.41	192
9*C*	52	2.05			1.99	0.86	21,300	48,100	0.47	187
9*D*	67	2.64			2.10	0.75	20,500	39,900	0.39	174
Group III. Cast at 1340°C. (2445°F.)										
7*AK*	22	0.86	2.82	1.63	1.63	1.19	......	51,200	0.37	248
7*BK*	37	1.46			1.78	1.04	32,300	62,000	0.45	212
7*CK*	52	2.05			1.83	0.99	28,200	58,800	0.47	207
7*DK*	67	2.64			1.85	0.97	29,200	57,100	0.49	196
8*AK*	22	0.86	2.84	1.86	1.73	1.11	......	45,400	0.30	255
8*BK*	37	1.46			1.83	1.01	30,600	63,600	0.55	214
8*CK*	52	2.05			1.90	0.94	27,300	56,500	0.47	196
8*DK*	67	2.64			1.94	0.90	25,300	50,800	0.45	187
9*AK*	22	0.86	2.85	2.23	1.76	1.09	......	52,600	0.41	248
9*BK*	37	1.46			1.80	1.05	29,500	61,500	0.49	202
9*CK*	52	2.05			1.90	0.95	25,600	54,900	0.51	192
9*DK*	67	2.64			2.02	0.83	25,500	49,700	0.45	183

TABLE 67.—RELATION BETWEEN PROPERTIES OF CAST IRON, COMPOSITION, CASTING TEMPERATURE, AND SIZE OF TEST BAR.*—(*Concluded*)

Number	Diameter		Composition, per cent				Tensile strength, lb. per sq. in.	Transverse test†		Brinell hardness
	mm.	in.	Total carbon	Si	Graphite	Combined carbon		Modulus of rupture, lb. per sq. in.	Deflection, in.	
Group IV. Cast at 1440°C. (2625°F.)										
10*A*	22	0.86	2.59	1.30	0.83	1.76	......	72,300‡	0.35	286
10*B*	37	1.46			1.40	1.19	36,000	57,800‡	0.28	241
10*C*	52	2.05			1.46	1.13	35,600	61,900	0.33	235
10*D*	67	2.64			1.53	1.06	32,200	60,100	0.35	228
11*A*	22	0.86	2.61	1.69	1.48	1.13	35,600	44,400‡	0.28	262
11*B*	37	1.46			1.55	1.06	35,900	66,900	0.37	235
11*C*	52	2.05			1.58	1.03	32,600	67,600	0.51	228
11*D*	67	2.64			1.70	0.91	28,100	58,200	0.43	217
12*A*	22	0.86	2.63	2.24	1.60	1.03	34,200	56,800‡	0.41	255
12*B*	37	1.46			1.70	0.93	30,000	61,900	0.43	207
12*C*	52	2.05			1.78	0.85	27,000	58,300	0.45	202
12*D*	67	2.64			1.86	0.77	26,500	53,800	0.43	187
Group IV. Cast at 1340°C. (2445°F.)										
10*AK*	22	0.86	2.59	1.30	0.68	1.91	......	79,500	0.33	
10*BK*	37	1.46			1.38	1.21	33,200	63,000‡	0.33	255
10*CK*	52	2.05			1.42	1.17	36,700	53,000‡	0.26	241
10*DK*	67	2.64			1.40	1.19	36,600	61,800‡	0.30	235
11*AK*	22	0.86	2.61	1.69	1.45	1.16	39,600	‡	....	262
11*BK*	37	1.46			1.48	1.13	34,600	61,800‡	0.37	241
11*CK*	52	2.05			1.58	1.03	32,600	63,900‡	0.45	231
11*DK*	67	2.64			1.60	1.01	28,500	60,500	0.43	220
12*AK*	22	0.86	2.63	2.24	1.58	1.05	......	....	‡	255
12*BK*	37	1.46			1.62	1.01	33,300	61,900	0.45	217
12*CK*	52	2.05			1.74	0.89	28,000	60,000	0.47	207
12*DK*	67	2.64			1.75	0.88	26,300	55,000	0.43	196

* Koch and Piwowarsky.[628]

† The bars 22 mm. (0.86 in.) in diameter were tested without removing the surface, the others were turned to a diameter of 30 mm. (1.2 in.); all were broken on a 600-mm. (23.6-in.) span.

‡ Flaw in test bar.

the carbon content decreases. These values also increase as the silicon content decreases. Figure 103 shows to somewhat better advantage how the modulus of rupture varies with the carbon content for iron containing 2.25 per cent silicon.

Most commercial cast irons contain from 0.4 to 1 per cent manganese;(582) varying the manganese content within this range has little influence on the properties of the iron if the silicon content is ample. It is considered that up to 0.15 per cent sulphur is not harmful in the majority of gray-iron castings.

The phosphorus content of American cast irons(582) ranges from 0.15 to 0.90 per cent; this element does not materially influence the strength of most types of irons. In high-test irons phosphorus is kept low since it tends to increase brittleness.

C. EFFECT OF SUPERHEATING AND LADLE ADDITIONS

The investigations of Elliott(51,66,83) and Saeger and Ash(653) in the United States, of Piwowarsky,(169,194) Wedemeyer,(203) Meyer,(223) Bardenheuer and Zeyen,(286) and Zeyen(502) in Germany, of Norbury and Morgan(398) in England, and of Tanimura(495) in Japan have shown that the structure and properties of cast iron are dependent upon the temperature to which the iron is heated prior to casting. It is generally acknowledged that Elliott's pioneering work in this field reported in 1919 to 1921 opened a new era in the production of high-strength cast iron.

111. Effect of Superheating on Mechanical Properties.—In general, as the temperature to which the molten iron is heated is raised the graphite flakes in the casting tend to decrease in size and the amount of combined carbon tends to increase. For some irons at least, as was indicated by Piwowarsky, the amount of combined carbon may first tend to increase as the melting temperature is raised and then, at high temperatures, tend to decrease. The time at which the iron is held at the highest temperature also influences the amount of combined carbon in the solid iron.

The data obtained by Meyer,(223) given in Table 68, indicate that superheating tends to improve the mechanical properties of some classes of gray cast iron. The iron on which the data were obtained contained 3.5 per cent carbon, 0.8 per cent silicon, 0.25 per cent phosphorus, and 0.1 per cent sulphur. Although superheating is now commonly used for the production of high-

strength irons, it has been quite definitely shown that superheating not only fails to increase the strength of certain irons but that it may actually lower their strength. Some of Bardenheuer and Zeyen's[(286)] data on the influence of superheating on the properties of several irons are given in Table 69. The irons were melted in an aluminum silicate crucible and were partly protected from the action of the atmosphere by a cover of aluminum silicate. Samples were cast into dry-sand molds. Microscopic examination showed that the size of the graphite flakes

TABLE 68.—INFLUENCE OF SUPERHEATING ON PROPERTIES OF A CAST IRON CONTAINING 3.5 PER CENT CARBON, 0.8 PER CENT SILICON, 0.25 PER CENT PHOSPHORUS, AND 0.1 PER CENT SULPHUR*

Heated to		Cast at		Tensile strength, lb. per sq. in.	Transverse test	
°C.	°F.	°C.	°F.		Modulus of rupture, lb. per sq. in.	Deflection,† in.
1250	2280	1250	2280	34,300	42,700	0.44
1300	2370	1250	2280	33,900	58,500	0.36
1350	2460	1250	2280	24,300	42,900	0.17
1450	2640	1250	2280	34,200	59,600	0.44
1530	2785	1250	2280	41,100	61,600	0.32

* Meyer.[(223)]
† Span = 20 *d*.

decreased as the temperature to which the iron was heated increased. However, as the temperature to which the low-carbon irons were heated was increased, the graphite tended to form a network which would obviously tend to weaken the iron. Although the data in Table 69 indicate that superheating had little influence on the strength of the iron, an analysis of data obtained from 24 melts showed that superheating tended to increase the strength of high-carbon irons containing normal amounts of silicon and manganese, but that it tended to decrease the strength of low-carbon irons (irons with 2.5 per cent carbon). The deleterious effect of superheating low-carbon irons could be avoided by increasing the silicon content of the iron or by adding powdered silicon to the ladle.

The influence of superheating on the properties of three types of cast iron (see page 268 for composition) was determined in

this country by Saeger and Ash,[653] who found that it did not greatly improve the properties of any of the irons. However, for an iron of 3.79 per cent total carbon, 1.32 per cent silicon, 0.73 per cent manganese, 0.12 per cent phosphorus, 0.06 per cent sulphur, recent work at the National Bureau of Standards[789] showed 7 to 8 per cent increase in transverse strength on bars of 0.75-, 1.20-, and 2.2-in. diameter, when the iron had been heated to 1600°C.

TABLE 69.—INFLUENCE OF SUPERHEATING ON PROPERTIES OF CAST IRON*

Melt, number	Composition, per cent						Heating temperature and time†	Casting temperature		Tensile strength, lb. per sq. in.	Transverse test‡		Brinell hardness
	C	Graphite	Si	Mn	P	S		°C.	°F.		Modulus of rupture, lb. per sq. in.	Deflection, in.	
5	3.37	2.58	1.50	0.91	0.45	0.077	*A*	1340	2445	34,500	61,100	0.41	199
	3.42	2.59	1.57	0.96			*B*	1330	2425	33,800	66,500	0.43	208
	3.42	2.60	1.58	1.07			*C*	1350	2460	34,900	69,200	0.34	211
6	3.20	2.46	2.02	1.02	0.33	0.087	*A*	1350	2460	34,900	64,900	0.49	211
	3.22	2.56	2.15	1.09			*B*	1350	2460	35,300	64,000	0.41	214
	3.26	2.59	2.06	1.08			*C*	1350	2460	34,500	57,800	0.28	215
7	2.59	1.81	2.28	0.93	0.35	0.066	*A*	1350	2460	41,000	82,500	0.45	224
	2.57	1.75	2.29	0.97			*B*	1350	2460	43,100	73,200	0.35	246
	2.61	1.83	2.28	1.01			*C*	1350	2460	46,700	70,600	0.36	243
8	2.30	1.59	2.47	0.99	0.37	0.057	*A*	1380	2515	34,500	80,800	0.36	240
	2.29	1.55	2.46	0.99			*B*	1380	2515	47,100	71,200	0.28	253
	2.30	1.47	2.49	0.97			*C*	1380	2515	48,500	66,600	0.31	254

* Bardenheuer and Zeyen.[286]

† *A* held 15 min. at 1350 to 1380°C. (2460 to 2515°F.); *B* held 15 min. at 1460 to 1500°C. (2660 to 2730°F.); *C* held 15 min. at 1580 to 1630°C. (2875 to 2965°F.).

‡ Bar of 30-mm. (1.18-in.) diameter broken on 600-mm. (23.6-in.) span.

(2910°F.), over that obtained when the iron had been heated only to 1500°C. (2730°F.). Superheating to 1700°C. (3090°F.) produced a further increase in transverse strength of 7 to 8 per cent over that from the superheating to 1600°C. (2910°F.).

112. Tanimura's Investigation of the Effect of Superheating on Properties.—Tanimura[495] made small laboratory heats, using fairly pure materials. His base was a white pig iron of 3.14 per cent total carbon, 0.15 per cent manganese, and only a trace of

silicon, with 0.046 per cent phosphorus and 0.032 per cent sulphur. Silicon additions were made with a ferrosilicon of 64.75 per cent silicon. Higher carbon melts were made by allowing the charge to pick up carbon from the graphite crucible, while low-carbon melts were made by dilution with Swedish wrought iron of 0.07 per cent carbon, 0.05 per cent manganese, 0.041 per cent sulphur, and 0.086 per cent phosphorus. The silicon in the wrought iron, most of which was present as slag, was 0.20 per cent.

Tanimura heated his melts for 10 min. at the predetermined temperature, which ranged from 1300 to 1600°C. (2370 to 2910°F.), and then cooled all the melts to the same temperature, 1250°C. (2280°F.), before casting. Some of the melts were poured into mercury to produce a chilled regulus, and others into a sand mold preheated to 500°C. (930°F.) to produce very slow cooling. The proportions of combined carbon and of graphite were then determined in order to show what effect superheating had upon the tendency to chill.

After heating alloys of the following composition

Element	Percentage			
	a	*b*	*c*	*d*
Carbon	3.21	3.16	2.98	2.99
Silicon	1.38	1.65	2.05	2.56

to temperatures of 1450 to 1525°C. (2640 to 2775°F.)—the higher silicon alloys requiring the higher temperatures—there was a larger proportion of combined carbon formed than when the alloys were heated at lower temperatures.

An iron with lower silicon (1.19 per cent) and 3.02 per cent carbon gave erratic results, the graphitic carbon in some samples being 50 to 60 per cent of the total carbon, and in others only 5 to 25 per cent, until the iron was heated to 1425°C. (2595°F.) when it became fully white. A similar iron, with 3.17 per cent carbon and 1.19 per cent silicon, poured into hot sand, did not show any chill until it had been heated to 1525°C. (2775°F.); but after heating at 1600°C. (2910°F.) it became fully white. Irons of the following composition

Element	Percentage			
	a	*b*	*c*	*d*
Carbon	3.60	3.12	2.80	2.91
Silicon	1.35	1.48	1.98	2.50

appeared to be unaffected by variations in melting temperature. In the slowly cooled samples and with still higher silicon and carbon contents, *viz.*,

Element	Percentage	
	a	*b*
Carbon	3.50	3.35
Silicon	2.07	2.89

superheating produced *more* graphite.

Tanimura concluded that the effect of superheating varies not only with the composition but also with the cooling rate. He

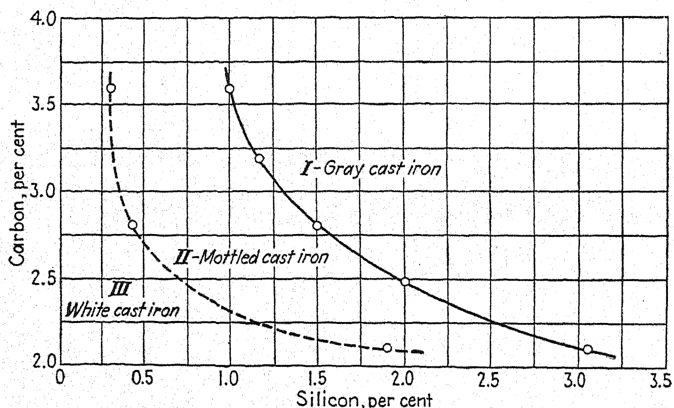

FIG. 104.—Composition limits for gray, mottled, and white cast irons. (*Tanimura.*[495])

plotted a structural diagram, Fig. 104, in which the compositions versus structures are given. Alloys whose composition lies in field I solidify as gray irons; when the composition is just to the right of the solid curve, they are high-strength gray irons. Alloys

whose composition lies in field II solidify as white irons when chilled and as gray irons when cooled slowly. Such alloys are suitable for chill castings. Alloys whose composition falls in field III are made gray with great difficulty, usually only by casting in a hot mold.

Regarding the effect of superheating, Tanimura found that this operation increased the graphite in irons whose composition is far to the right of the solid curve in field I; superheating had little effect on those alloys in field I close to the solid curve and induced chilling in those alloys whose composition lies within fields II and III. When the alloys are chill cast, superheating the irons whose composition lies just to the right of the solid curve in field I favors the formation of a white fracture. Superheating alloys whose composition lies farther to the right in field I may also result in white iron; but as the distance from the solid curve into field I increases, higher superheating temperatures are necessary to produce the white fracture. Tanimura then made tensile tests on superheated irons of 2.8 per cent and 3.6 per cent carbon with varying amounts of silicon and concluded that the highest strengths were obtained with an iron which, for a given carbon content, has as little silicon as will keep it gray—*i.e.*, has a composition not far to the right of the solid line in Fig. 104. Higher silicon tends to lower the strength. In choosing the carbon content for high-strength irons which are superheated, the lower carbon contents are preferred since only in these is the graphite fine instead of coarse, provided, of course, that the silicon is high enough. This is brought out in Fig. 105. The highest strength is obtained in an iron with 2.8 per cent carbon and 1.0 to 2.0 per cent silicon. With higher silicon percentages the strength is slightly lower, and with less than about 1 per cent silicon the strength drops rapidly. In the higher carbon iron there is a very narrow range of silicon—1.0 to 1.5 per cent—where the strength is fairly high. Increasing the silicon content of the 3.6 per cent carbon iron has a more marked effect on the tensile strength than a corresponding increase in the silicon of the 2.8 per cent carbon alloy.

113. Effect of Ladle Additions on Mechanical Properties.—A method of treating an iron of 2.8 per cent carbon with such low silicon that it would be on the dividing line between a pearlitic and a mottled iron, as indicated by Fig. 97, is given by Smal-

ley.[343] He added calcium silicide in the ladle, in the proportion of 120 oz. per ton, to an iron of 2.8 per cent carbon, 1.2 per cent silicon, 0.6 per cent manganese, 0.1 per cent sulphur, and 0.18 per cent phosphorus and found that small castings which would have been white without the addition had finely distributed, small graphite flakes in a pearlitic matrix. Standard test specimens of this iron, so treated, were stated to have a tensile

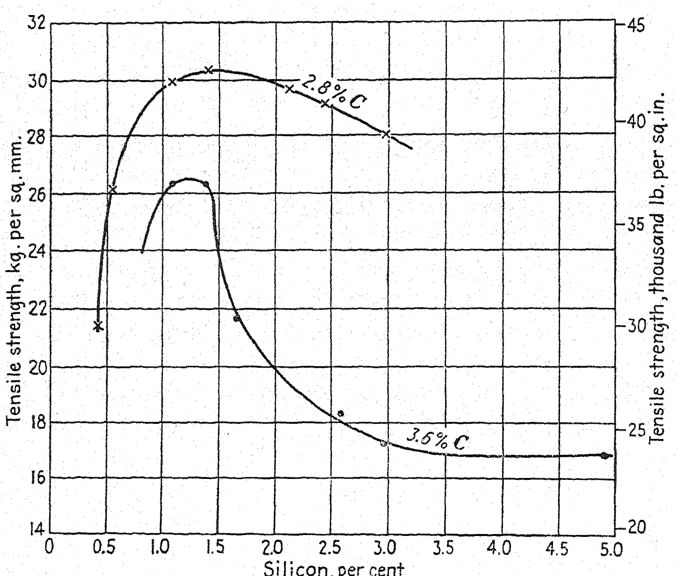

FIG. 105.—Relation between silicon content and tensile strength of superheated low- and high-carbon cast irons. (*Tanimura.*[495])

strength of 47,000 to 51,000 lb. per sq. in. with a perceptible elongation of 0.5 to 1 per cent, a Brinell hardness of 225 to 250, and a transverse strength ($1\frac{1}{8}$-in. diameter bar, 12 in. between supports) of 4800 lb. with about 0.13 in. deflection. A similar iron, of the same carbon content but with 1.65 per cent silicon, treated with 100 oz. of calcium silicide per ton, showed 46,000 lb. per sq. in. tensile strength, 0.5 per cent elongation, 255 Brinell, and 4300 lb. transverse strength with 0.16 in. deflection.

Lemoine[732] mentioned the addition of silicochromium or silicotitanium alloys to the ladle for purposes similar to those sought by Smalley in the use of calcium silicide, but stated that a ladle addition of silicon in the form of ferrosilicon containing 90

per cent silicon is also effective. He found that the excessive shrinkage and porosity met with in superheated cast iron containing 2.50 per cent carbon or less and 2.50 per cent silicon can be avoided by raising the carbon to 2.55 to 2.70 per cent and lowering the silicon to about 2.00 to 2.20 per cent; such charges melted in the electric furnace had tensile strengths of 41,000 to 49,000 lb. per sq. in. on bars cut from 3-in. sections of actual castings. Such iron, made in the cupola with steel charges, was however found to lack fluidity and to show too much tendency to chill. Addition of part of the silicon to the ladle was found to be helpful in remedying these troubles.

Lemoine concluded that instead of making the melt direct from steel scrap it is preferable to make first, in the regular cupola, a low-carbon pig iron of 2.20 per cent or less carbon, 1.50 to 2.00 per cent silicon, and 1.20 to 1.50 per cent manganese and to treat this molten charge with soda ash to remove silicate inclusions which are thought to injure the fluidity of the metal. Some 40 to 50 per cent of this pig is then remelted with 15 to 20 per cent shop scrap and 30 to 35 per cent steel scrap and treated in the ladle with ferrosilicon (90 per cent silicon). R. S. MacPherran* commented that this practice was in use in 1895 and is still in daily use in Great Britain.

Coyle[207] gave the diagram shown in Fig. 106 and later Coyle and Houston[291] discussed methods for producing the low-carbon irons in the region of highest strength. At least 75 per cent steel scrap with 25 per cent cast-iron scrap is charged together with ferrosilicon and ferromanganese so that the metal tapped from the cupola contains from 2.50 to 3.10 per cent carbon with not more than 0.75 per cent silicon and between 0.50 to 0.80 per cent manganese. If such metal were cast directly, the castings would be white. An addition of ferrosilicon (or ferrosilicon and nickel) is made as the metal flows into the ladle to adjust the content of graphitizer. The graphite is then precipitated in very fine, evenly distributed particles. The metal should be cast promptly, or the graphite may coalesce into large flakes. Coyle and Houston discussed the process chiefly in relation to the use of nickel in metal so melted, and the detailed data they cited are for nickel irons. They stated, however, that without nickel tensile strengths up to 49,000 lb. per sq. in. can be obtained

* Private communication.

in plain iron cast in arbitration bars, but that the center of 3-in. rounds is weaker.

MacPherran[319, 546] described a similar process, consisting of melting a charge containing 95 per cent steel scrap, 4.5 per cent ferrosilicon (50 per cent silicon), and 0.5 per cent ferromanganese (80 per cent manganese) and making a final addition at the spout of 6 lb. per ton of silicomanganese containing 50 per cent silicon and 25 per cent manganese. While additional alloys are often used to secure still higher strength, MacPherran pointed out that the use of such alloys is not essential to the production of

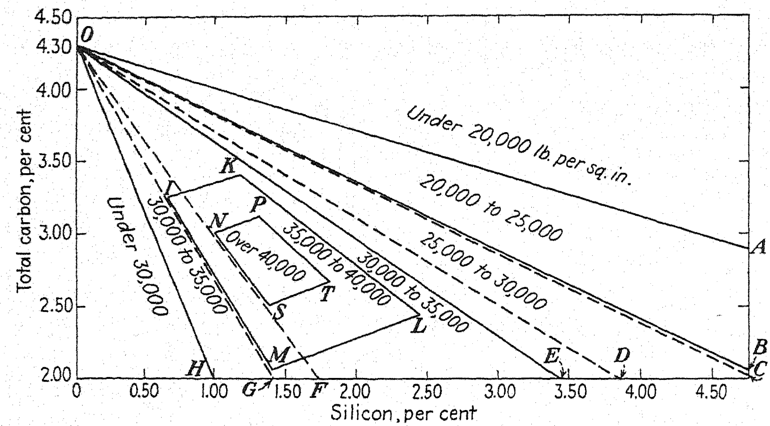

FIG. 106.—Cast-iron diagram showing tensile strength versus composition. (*Coyle.*[207])

high-strength cast iron. He stated that an iron with 2.5 to 3.0 per cent total carbon, 1.9 to 2.65 per cent silicon, 0.7 to 0.8 per cent manganese, about 0.1 per cent sulphur, and from a trace to 0.2 per cent of residual nickel had tensile strengths of 51,000 to 62,000 lb. per sq. in.

114. Explanations of the Effect of Superheating on Properties.—The most popular explanation of the influence of the temperature reached in melting on the structure of the solid iron is that graphite nuclei are present in the molten iron and that they gradually disappear as the temperature is raised or as the time at temperature is increased. Piwowarsky[194] considered the graphite-nuclei theory as a possible explanation of the behavior up to the temperature of reversal, and this explanation became general. He himself, however, challenged this conclusion in a

later article and presented observations to prove that graphite nuclei could not be present in iron heated to just above the liquidus.[648] In his most recent report on graphitization Piwowarsky[812] stated that even coarse graphite dissolves so rapidly in molten cast iron that graphite nuclei can play no part in graphitization. This has been discussed in Volume I, Chapter XII.

The reversal by which the highest melting temperatures appear to produce a recurrence of coarse graphite, *i.e.*, failure to undercool, has been chiefly noted in laboratory melts. Bardenheuer

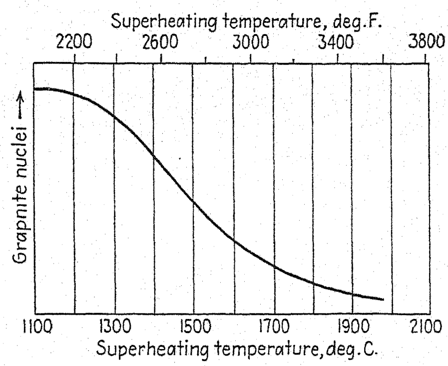

FIG. 107.—Relation between superheating temperature and number of graphite nuclei. (*von Schwarz.*[655])

and Zeyen[248] ascribed this to attack upon the crucible wall and showed a photograph of a melt frozen in the crucible from a maximum temperature of 1800°C. (3270°F.) in which coarse graphite separated at the periphery of the melt, close to the crucible, indicating contamination. With such disturbing factors omitted, von Schwarz[655] considered that the correct curve for the effect of overheating on existence of graphite nuclei, or the phenomenon ascribed to it, is as shown in Fig. 107.

As mentioned in Volume I, Chapter XII, von Keil and coworkers[725] postulated the existence of ferrous silicate particles as an alternative to the graphite-nuclei theory. Lemoine[732] has already been cited (page 317) as believing that silicates may be present in amounts sufficient to affect fluidity, and that part of them may be removed by soda ash.

Walther[782] found that separation of the desired type of graphite nuclei in high-test iron may be secured by addition to

the electric furnace, just prior to the final silicon additions, of scrap, of the same analysis as the bath, which has its graphite in the desired fine, nodular form, while, if coarser graphite and hence better machinability is wanted, the scrap should have coarse graphite present.

As a result of experiments with vacuum melting and with deoxidation by aluminum, Köttgen(385) concluded that the effect of superheating of the melt upon the size of graphite flakes is connected with gas content. Lemoine(732) emphasized the marked difference in the effect of silicon added early in the melt compared to that added just before pouring, and this recalls the discussion on "overreduced" steel castings where the time of addition of silicon appears to affect the gas content. Probably a number of interrelated phenomena occur in a melt of cast iron, and which one is of primary importance in controlling the type of graphite formed remains to be shown.

Piwowarsky(752) listed among the factors which influence the size and distribution of graphite, superheating, stirring of the bath, time of addition of silicon, carbon content, and content of alloying elements. Separation of gas from the melt is mentioned as perhaps having an effect similar to that of stirring the bath. The presence of slag low in oxide tends toward the production of fine graphite. Piwowarsky again remarked (see page 318) that the idea of undissolved graphite flakes in the melt seems no longer tenable and that something else in the melt must be responsible for the phenomena so often ascribed to them. Molecular groupings of carbides, silicides, etc., in the liquid state, which are more or less dissociated under different conditions, were suggested as a possibility.

The effect of slag, mentioned by Piwowarsky, has been discussed in Volume I, Chapter XII, in its relation to von Keil's "silicate slime" idea. W. Ruff* suggested introducing oxide particles in the finishing stage of a cast-iron heat, to serve as nuclei for graphitization, instead of adding ferrosilicon or calcium silicide. He claimed that silicon alone, in the absence of oxygen, does not cause graphitization, and that white iron can be obtained with a silicon content over 1.7 per cent, provided that oxygen is not present to form nuclei for graphite separation. He suggested heating in a reducing atmosphere (superheating is said to aid,

* U.S. Patent 1,985,553, Dec. 25, 1934.

but it is to be avoided if possible) and removing the oxygen-containing slag, then adjusting the silicon content (a minimum of 0.5 per cent silicon is required), and finally oxidizing to a controlled degree by use of an oxidizing atmosphere, an oxidizing slag, or addition of iron or manganese oxide. By this regulation of the oxide content before pouring, the degree of undercooling and the number, distribution, and size of graphite particles are said to be under control.

It is alleged that white iron can be thus taken direct from the blast furnace, adjusted for silicon content, then given the controlled oxidizing treatment, and finally annealed several hours at 675 to 735°C. (1245 to 1355°F.) to break down pearlite. By securing a very fine structure in the graphite formed during solidification and applying this short low-temperature heat treatment material comparable to malleable iron is said to be produced.

The comments of Köttgen and of Piwowarsky (see page 320) that gas may have something to do with the type of graphite produced are apparently about to be justified, for, according to a preliminary announcement by the British Cast Iron Research Association,[(681)] Norbury has found it possible to change the structure from one type to another by treatment with suitable gases and has evolved a method by which a suitable addition to the metal and a simple aftertreatment of the molten metal produce, on a laboratory scale at least, the finely divided undercooled form of graphite, regardless of the original total carbon. It is also stated that this can be done in heavy sections, and that the action can be reversed and flake graphite produced from a melt that would give fine graphite. Details had not been published at the time of preparing this volume.

Comstock,[(597)] Paschke and Schuster,[(747)] and others have pointed out the effect of small amounts of titanium in favoring the production of fine graphite.

Girardet[(297)] has described a swirling fore-hearth into which cupola metal is tapped and in which it is whirled for a minute or two. He stated that in an ordinary iron of 2 per cent silicon, 1.3 per cent phosphorus, and 0.4 per cent manganese (carbon and sulphur contents not given) whirling produced finer graphite, raised the Brinell number from 145 to 160, and improved the Frémont impact-test result about 40 per cent. The claim was

made that, once the modified graphite form has been obtained, "heredity" will preserve it on a remelt. The action is ascribed, among other things, to elimination of gas and modification of liquid emulsified materials.

Girardet* stated in 1934 that the swirling apparatus, somewhat modified from its first form, has operated satisfactorily for 4 years, but that no other installations have ever been made. He stressed the point that a type of stirring action is required which will give an area of relative quietude so that gas may be removed and slag particles in suspension may be agglomerated, rather than a type of stirring which will tend to emulsify the slag particles in the metal. Irresberger[182] had found earlier that agitation was helpful; and the use of a rocking electric furnace in superheating cast iron to produce higher strength and finer graphite than is obtained in unstirred melts of the same composition has also been reported.[603]

Nipper[744] pointed out that upon superheating, impurities, consisting of coarse or more or less dispersed oxides and silicates, coagulate and separate out. The high degree of purity thus obtained allows undercooling which produces fine graphite. Pearce, in discussion of Nipper's paper,[744] remarked that English investigators have regarded superheating as producing an undercooling which in turn produces graphite so fine as to be undesirable, and later siliconization as serving to correct this and produce graphite in a fine but not too fine state of division. The Germans, on the contrary, seem to consider that undercooling is promoted by the final addition of silicon. Pearce suspected that each may be right in the case of the pig irons he is using, and that the investigators are "converging on the truth."

Thus the factors involved in control of size and distribution of graphite appear to be explainable without resorting to an assumption that actual graphite nuclei are present or absent in the melt. If it is, instead, assumed that non-metallic inclusions or tiny gas bubbles, released as the metal freezes, are the nuclei involved, the control of the nuclei by superheating, slagging, stirring, control of atmosphere, and by the introduction of small amounts of metallic elements which readily form oxides, nitrides, or stable carbides, becomes explainable much as the old idea of "body" in steel is explainable by the theory of submicroscopic

* Private communication to Dr. G. H. Clamer.

particles which produce small grain size. If for grain size in steel graphite size in cast iron is substituted, the theories become very similar. "Finishing" the heat of cast iron along lines analogous to finishing the heat of steel seems to be what modern practice is tending to do. Obviously the possibility of controlled finishing in the cupola itself, or even in the ladle, is limited, and other methods of melting or of final treatment in a duplexing operation offer greater possibilities.

Although much emphasis has been laid on high-strength iron and on methods for control of the size of the graphite particles in low-carbon irons, a comment of Pearce(749) is worth repeating:

> A word of caution is required about specifying these high-duty irons. They are more difficult to cast, for they have a high melting temperature, a shorter fluid range, and greater shrinkage than ordinary cast irons, and it does not follow that any and every casting can be made in them. The necessity for securing varied thicknesses in one piece forms the very *raison d'être* of the average casting, and the design and size often compel the use of a composition permitting a greater degree of castability.

D. EFFECT OF OTHER VARIABLES ON MECHANICAL PROPERTIES

There are a few variables met with in the production of iron castings, in addition to the composition and the operations described under the term superheating and ladle additions, which affect the mechanical properties. The most important of these are the casting temperature, the size of the section cast, and the casting skin. Data on these variables are summarized below. There is also some information available on iron as cast centrifugally and as cast into permanent molds, and on the properties of white and chilled cast irons, which are likewise summarized below.

115. Effect of Casting Temperature on Properties.—The properties of gray cast iron are naturally somewhat influenced by the casting temperature employed, but they are more nearly constant for different pouring temperatures than is the case with many other alloys. Bardenheuer and Zeyen(248) found that for an iron containing 3.6 per cent carbon, 1.82 per cent silicon, 0.69 per cent manganese, 0.49 per cent phosphorus, and 0.116 per cent sulphur, which had been heated to 1400°C. (2550°F.), the

modulus of rupture and the tensile strength varied with the casting temperature* as follows:

Casting temperature		Modulus of rupture, lb. per sq. in.	Tensile strength, lb. per sq. in.
°C.	°F.		
1380	2515	47,200	20,900
1340	2445	52,100	23,600
1305	2380	50,900	22,900
1270	2320	46,800	23,000
1230	2245	46,400	24,300
1180	2155	44,700	23,500

116. Influence of Size of Section on Properties.—The rate at which a casting cools decreases as the size of the casting increases. Since the properties of cast iron depend on the rate of cooling, they also depend on the mass of the casting. The cooling rate is not constant throughout a casting; portions near the surface cool more rapidly than portions near the center, and the strength of specimens taken near the surface may be appreciably greater than that of sections near the center or axis. Consequently, for large sections, specimens taken from near the axis may not be at all representative of the over-all strength of the material. This is brought out in the Symposium on Cast Iron(582) where data are presented to show that as the diameter of the cast test bar increases the percentage decrease in modulus of rupture is much less than the percentage decrease in tensile strength of specimens cut from the axes of the cast bars. Data reported by Pearce(555) proved that the decrease in strength as size of section increased did not result from a decrease in the amount of combined carbon. The size of the graphite flakes, however, increased as the size of the section increased.

Figures 102 and 103 (pages 305 and 306) show the decrease in strength with increase in diameter of cast bar for several irons.

* MacKenzie (private communication) in commenting on this portion of the manuscript called attention to the fact that the iron used by Bardenheuer and Zeyen was of an analysis which is insensitive to variations in casting temperature. According to MacKenzie, if they had used a low-carbon iron, the effect of casting temperature would have been more pronounced.

The data reported by Bornstein,[438] given in Table 70, provide a somewhat better picture of the decrease in strength as the diameter of the cast bar increases.

It is well known that the cooling rate and, consequently, the structure and properties of cast iron are to a considerable degree dependent on the size of the section. Thus it is possible for the corners or surfaces of a casting to chill while the interior cools slowly enough during freezing for coarse graphite to separate.

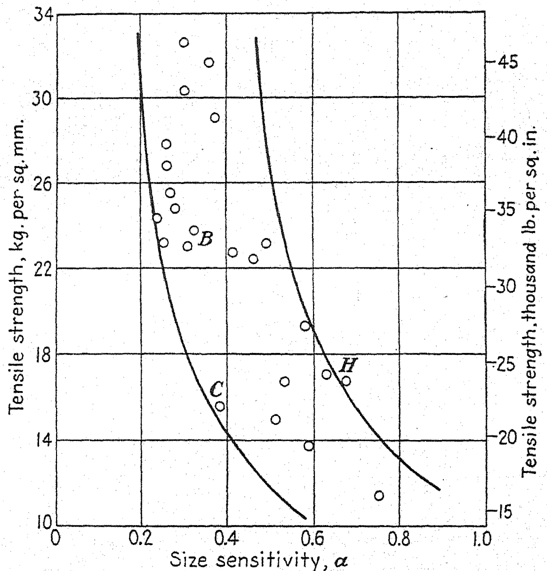

FIG. 108.—Relation between tensile strength and size sensitivity. (*Heller and Jungbluth.*[715])

In general, according to Heller and Jungbluth,[715] the effect of section size is less pronounced with the stronger irons. These investigators worked out the relation between tensile strength and section size as shown in Fig. 108. The "section sensitivity" factor a is the exponent in the equation

$$\frac{S_d}{S_{30}} = \left(\frac{d}{30}\right)^a$$

where S_{30} is the tensile strength determined on the standard 30-mm. diameter bar, S_d the tensile strength determined on a bar of another diameter, and d is that diameter in millimeters.

The exponential nature of the sensitivity relation is brought out by Heller and Jungbluth in Fig. 109 for the irons listed in Table 71. It should be noted that the ordinate of Fig. 109 is in percentage, the strength of each iron in the 30-mm. diameter specimen being taken as 100 per cent. As will be seen in Table 71, the strengths vary widely. The slope of each curve in the

TABLE 70.—RELATION OF TENSILE STRENGTH AND MODULUS OF RUPTURE TO DIAMETER OF CASTING*

Tensile strength, lb. per sq. in.				Modulus of rupture, lb. per sq. in.			
1.2-in. diameter bar, as cast	Percentage, as compared with tensile strength of 1.2-in. diameter bar			1.2-in. diameter bar, as cast	Percentage, as compared with modulus of rupture of 1.2-in. diameter bar		
	0.75-in. diameter bar	2-in. diameter bar	3-in. diameter bar		0.75-in. diameter bar	2-in. diameter bar	3-in. diameter bar
Soft iron							
23,600	122	65	51	41,800	133	94	78
25,400	107	71	57	43,000	121	91	76
24,500†	115†	68†	54†	42,400†	127†	93†	77†
Machinery iron							
35,200	104	69	53	58,400	155	95	80
36,000	104	77	63	60,900	113	100	94
36,900	101	84	66	65,000	117	97	82
36,000†	103†	77†	61†	61,400†	115†	97†	85†

* Bornstein.(438,582)
† Average.

logarithmic plot shows the exponent a. Of course, the lines would not be straight if a structural change occurred, *i.e.*, they are valid only for gray iron and would show a break when the iron starts to become mottled.

Rother,(136) Rother and Mazurie,(195,335) the American Society for Testing Materials,(245) Pearce,(277) Bolton,(249,288) MacPherran,(318) Piwowarsky and Söhnchen,(650) and von Schwarz and Väth(656) have all dealt with this general topic. It is further

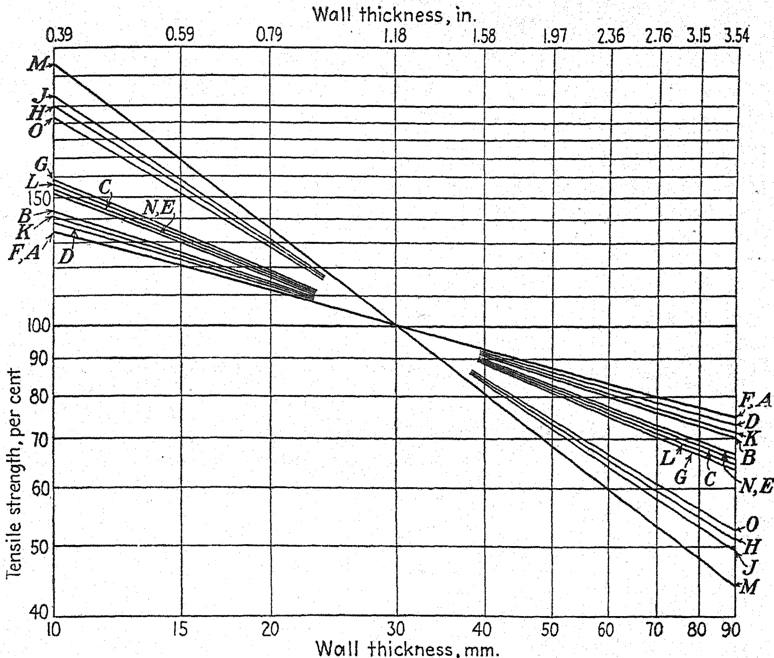

FIG. 109.—Relation between tensile strength and wall thickness. (*Heller and Jungbluth.*[715])

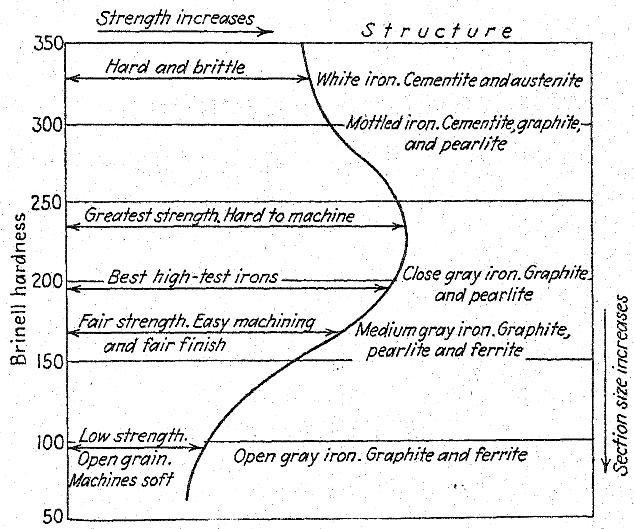

FIG. 110.—Relation between structure, section size, and properties of cast iron. (*Symposium on Cast Iron*[582] *after Bolton and Bornstein.*)

illustrated in Fig. 110, taken from the Symposium on Cast Iron,[582] where the effect of change in structure is emphasized.

117. The Casting Skin.—Since the properties of the interior of a gray-iron casting are poorer toward the center, the relative properties of a machined and an unmachined test bar will depend on the mold surface and on the way the particular iron behaves at different cooling rates. That Meyersberg's[552] findings about the small effect of removing the skin on modulus of rupture

TABLE 71.—IRONS STUDIED BY HELLER AND JUNGBLUTH[715]

Iron	Composition, per cent					Tensile strength, lb. per sq. in.
	C	Si	Mn	S	P	
A	3.29	1.22	0.50	0.11	0.38	38,000
B	3.28	1.79	0.49	0.10	0.35	33,000
C	3.70	1.72	0.33	0.10	0.11	22,000
D	3.23	1.60	0.68	0.10	0.44	35,500
E	3.24	1.28	1.02	0.10	0.25	45,000
F	3.22	1.50	0.84	0.10	0.18	39,500
G	3.17	2.44	0.67	0.09	0.15	32,000
H	3.43	2.43	0.89	0.07	0.09	25,000
J	3.66	1.95	0.65	0.095	0.077	24,000
K	3.20	1.49	0.51	0.12	0.24	43,000
L	3.49	1.87	1.08	0.10	0.25	32,500
M	3.78	2.19	0.52	0.07	0.07	16,000
N	3.18	1.45	0.85	0.09	0.13	41,000
O	3.54	2.09	0.44	0.09	0.07	27,500

(see page 290) will not hold for the comparison of specimens from the edge and the center of a large section of iron susceptible to the size effect is clear from the above discussion. The properties of the very outer skin of a casting may also vary from those of the underlying metal. Diepschlag[687] found that the surface of gray iron is lower in carbon and higher in silicon and sulphur than the core. The results of individual tests varied somewhat, and his data hardly serve to give general quantitative figures for loss of carbon by oxidation and for separation of other elements, but they substantiate Diepschlag's statement that the skin does not have the normal composition of gray iron.*

* According to MacKenzie (private communication), this depends largely on the mold facing.

Krynitsky and Harrison[386] studied this matter in particular relation to the blistering of cast iron on enameling and found that the carbon was lower at the surface, the sulphur slightly higher, while the silicon (contrary to Diepschlag's results) was somewhat lower. They noted an increase in combined carbon at the surface, and the "microchilled" layer was considered to be responsible for blistering. The surface was also found appreciably higher in nitrogen than the interior. For most uses of cast iron the variation in composition and properties at the surface is without effect, but in cases like enameling the variation may be of prime importance.

Comparisons of transverse bars cast in 0.875-, 1.20-, and 2.0-in. diameters, of 0.505-, 0.80-, and 1.25-in. diameter tensile specimens machined from them, and of separately cast tensile specimens 1.10 and 1.80 in. in diameter, are contained in a recent report of the A.S.T.M. Special Subcommittee on Test Bars.[677] This covers both plain and alloy cast irons but does not complete the Subcommittee's program, which will probably lead to some alterations in the recommendations for the preparation of test bars and to revisions in the thickness of sections to be represented by test bars of various sizes. The effect of size is brought out in the tables given in the report. One high-strength iron (a steel mixture without an alloy, composition not stated) gave 54,000 lb. per sq. in. tensile strength on the 0.505-in. diameter bar machined from an 0.875-in. diameter transverse bar, and 38,000 on the 1.25-in. diameter bar machined from a 2-in. diameter transverse bar. An iron of 3.58 per cent total carbon, 2.23 per cent silicon, 0.083 per cent sulphur, 0.46 per cent phosphorus, and 0.58 per cent manganese gave 30,000 and 17,000 lb, per sq. in. tensile strength in the two sizes mentioned above.

118. Centrifugally Cast Iron: The Sand-mold Process.—Cast iron is, of course, normally poured into green-sand molds, often with dry-sand cores. Most of the published information on the effects of composition, size of section, superheating, and other variables discussed in the previous pages refers to the rate of cooling obtained in a sand mold.

Since cast iron is a suitable material for water mains and the like, and as a large tonnage of cast-iron pipe is produced, such symmetrical cylindrical objects can be economically cast centrif-

ugally into rotating molds. A specialized industry for the centrifugal casting of iron pipe has therefore arisen. Some application is also being made of this process in the production of cylinder liners for internal-combustion engines, for locomotive piston-valve liners, and for brake drums.

There are two main processes for centrifugal casting. The first is the sand-mold process, described in detail by MacKenzie;[217] the other is the metal-mold or deLavaud process, described by Capron.[206] Since these two descriptive papers were published in 1927, a fairly extensive literature on the manufacture of centrifugally cast iron has grown up. Most of this literature deals with plant installations and with improvements in equipment and technique. Some of the recent improvements (as an example may be cited the use of a thin refractory inner lining in the metal-mold process, noted briefly below) have had a marked effect on the properties of the centrifugally cast pipe, but, unfortunately, extensive data are as yet unpublished. Therefore, as most of the published literature dealing with properties of centrifugally cast pipe does not represent the recent improvements in the process, the discussion below is restricted chiefly to a brief outline of the details of manufacture.

The sand-mold process consists basically of pouring molten iron of carefully controlled composition into a slowly rotating green-sand mold which is inclined at an angle to the horizontal. As the pouring continues, the mold is slowly lowered to the horizontal, and shortly before or just at the time it reaches this position, when all of the metal has been poured into the mold, the speed of rotation is suddenly accelerated to that required for the specific diameter and wall thickness of the pipe. Rotation is continued until solidification is complete, after which the mold is stripped from the pipe and the latter cleaned, inspected, coated, and tested.

According to MacKenzie,[217] sand-cast centrifugal pipe has an unusually good surface, free from defects and blemishes, and a smooth defect-free inner wall. Pipe cast by this process is uniform in section and in microstructure and is free from strains.

The fracture of the iron shows an extraordinarily clean, fine, and even grain. The slow cooling from outside surface to inside surface, while under the high pressure of centrifugal force, brings to the inside surface all foreign particles such as sand, blacking, or slag, and there is con-

siderable migration of manganese sulphide in the same way. Oxides or silicates of iron and manganese are also forced out.

Adding a little soda ash to the molten metal in the ladle produces a small amount of liquid slag which carries the above mentioned impurities with it to the inside wall of the solidified pipe, from which it is readily removed.

Microscopically, the iron cast by this method is, as in pipe cast in stationary sand molds, predominantly pearlitic. Combined carbon averages 0.75 per cent, silicon 1.6 per cent, and total carbon 3.5 per cent. The graphite particles, upon which the properties depend to a considerable degree, are in the form of fine whorls which grow slightly larger near the inner wall. MacKenzie quoted test data obtained by Talbot and Richart[201] which indicate that pipe centrifugally cast by the sand-mold process has markedly better properties by a variety of tests than the corresponding pipe cast in stationary sand molds. Pipe cast by this process has, according to MacKenzie, the advantage of evenness and soundness of section together with the true gray-iron structure, resilience, machinability, and corrosion resistance of pipe cast in stationary sand molds.

119. Centrifugally Cast Iron: The Metal-mold Process.—In the metal-mold process,[206] iron of carefully controlled composition is cast into a rotating water-cooled steel mold inclined at a slight angle to the horizontal. Pouring takes less than a minute, solidification is complete in a few minutes more, and the pipe is immediately withdrawn from the mold. The time for casting and solidification is so short that only a slight surface chill and minor casting strains result. These are removed by an immediate annealing just below the critical temperature. As cast by this process, the section is uniform, the graphite particles are small and well distributed, and the strength is high. One of the great advantages of the metal-mold process—which is true for other centrifugal processes as well—is the reduction in cost. The metal-mold process is the oldest process for the centrifugal casting of ferrous products, having been used on a commercial scale since 1922.

In order to decrease the depth of chill and produce pipe with a pearlitic gray-iron body and an outer surface only slightly chilled, actual contact of the casting with the metal mold is

avoided by blowing in a very thin coating of dry, finely pulverized material which is held by the centrifugal action of the rotating mold in a film about 0.001 in. thick. The film may be an inert material like kaolin, or it may be one like pulverized ferrosilicon which, if fused into the periphery, will provide added silicon at that point to reduce the chilling effect [756] (U. S. Patent 1,954,892, April 17, 1934). By this means it is said that in irons of 3.5 to 3.7 per cent total carbon, 1.7 to 2.2 per cent silicon, about 0.7 per cent phosphorus, and normal manganese and sulphur contents the periphery is made to contain less than 0.2 per cent combined carbon although it is practically free from coarse graphite, while the body of the pipe will contain 0.3 to 0.5 per cent combined carbon. Annealing is still carried on, but at a lower temperature and for a longer time than with metal-mold pipe cast without the facing. Pipe made with the facing has much better impact resistance than that made without it. The graphite in such castings and even in those not subsequently annealed may approach the "temper" form.

Most of the published information on "metal-contact" (deLavaud type) pipe refers to material which has been annealed to break down some of the original combined carbon. Examples of this type of pipe, after annealing, are shown by Talbot and Richart[201] to contain 3.5 to 3.6 per cent total carbon, about 0.1 per cent combined carbon, 1.65 to 2 per cent silicon, 0.8 per cent phosphorus, and normal manganese and sulphur.

Main[547] has discussed the structures which accompany acceptable performance in centrifugally cast automobile brake drums. Cast iron alloyed with nickel, with nickel and chromium, or with molybdenum has been used more than ordinary cast iron.

Hurst[181,378,457] described what he calls "spun sorbitic" cast iron, a low total-carbon, low-silicon iron which would normally cool with a pearlitic structure. By playing an air blast on the inner surface of the cooling centrifugal casting the metal may be given a somewhat sorbitic structure, said to be highly wear resistant.

Several publications[134,177,780] deal with the possible segregation in centrifugal cast iron during freezing. That manganese sulphide tends to accumulate toward the inside of the casting is evidence that it is not very soluble in molten cast iron.

No very marked segregation of other elements has been noted.

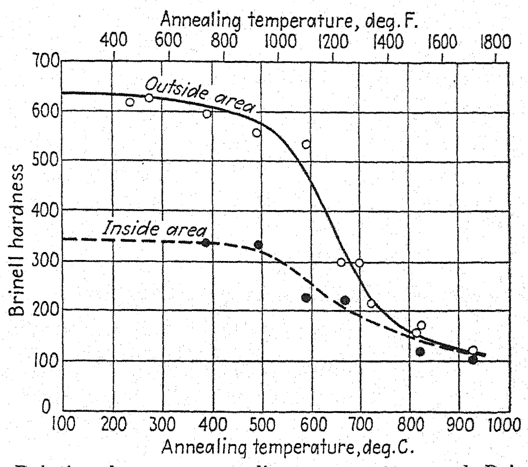

FIG. 111.—Relation between annealing temperature and Brinell hardness of centrifugally cast pipe, annealed 4 hr. Wall thickness 5 mm. (0.2 in.). (*von Schwarz and Väth.*[412])

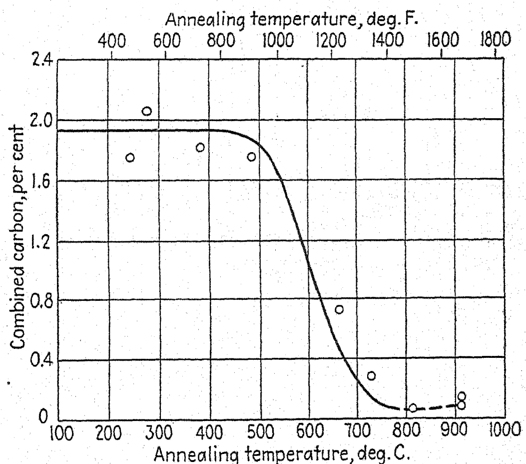

FIG. 112.—Relation between annealing temperature and combined carbon of centrifugally cast pipe, annealed 4 hr. Wall thickness 5 mm. (0.2 in.). (*von Schwarz and Väth.*[412])

Pardun[134] mentioned a 30-min. anneal at 850 to 900°C. (1560 to 1650°F.) as suitable for annealing metal-mold cast pipe. Von Schwarz and Väth's[412] Figs. 111 and 112 show the effect

of the annealing process on hardness and on the amount of combined carbon.

120. Iron Cast in Permanent Molds.—The problem of casting gray iron into permanent metal molds or metal molds lined with a thin refractory layer to reduce the chilling effect is quite similar to that of centrifugal casting. A soft iron which does not chill readily is required, and in some cases, especially if high machining rates are used, a short anneal is necessary to break down the chilled layer if one is formed. The speed of cooling, especially in castings of small section, which is dependent upon the pouring temperature, mold temperature, and the mass and thermal diffusivity of the mold, must be very accurately controlled.

Very material improvement in strength, without prohibitive decrease in machinability, results from the fine grain and the small evenly distributed graphite flakes produced by the rapid cooling. Another advantage of the process is that the castings have a clean, sand-free surface. Casting gray iron in permanent molds has been described repeatedly.*

121. Mechanical Properties of White and Chilled Cast Irons.—The manufacture of white iron and the method of chilling gray iron to produce a hard surface on those parts of the casting requiring it have been described briefly in the previous chapter (page 272). As noted there, white cast irons are not readily classified on the basis of properties as the structure and its accompanying properties may vary widely from casting to casting and from one section to another in the same casting. There are consequently few data available on the mechanical properties of white and chilled irons. Most of these have been summarized in the recent Symposium[582] from which Table 72 is taken.

Within certain limits, the hardness and brittleness of white cast iron may be controlled by regulating the carbon content; as the carbon increases, hardness and brittleness both increase. According to results reported in the Symposium on Cast Iron,[582] the scleroscope hardness of a white iron containing 3.10 per cent carbon was 63; it increased to 75 for a carbon content of 3.90 per cent.

* For articles on the subject, see bibliography, reference Nos. 43, 50, 52, 185, 190, 196, 236, 266, 285, 354, 385, 518, and 659.

Pohl,[813] whose work on the effect of variations in carbon, and on the effect of temperature and time of heating before pouring, on depth of chill was reviewed on page 272, also determined the effect of carbon on the scleroscope hardness of chilled iron. As the carbon increased from 2.5 to 4.1 per cent the scleroscope hardness increased linearly from 55 to 82; the depth of chill also

TABLE 72.—TYPICAL RESULTS OF TESTS ON CHILLED PLAIN IRONS*

Composition and property	Iron	
	High carbon	Low carbon
Total carbon, per cent	3.50	2.75
Silicon, per cent	0.75	0.75
Brinell-hardness number of chilled surface	500	400
Tensile strength in chilled section, lb. per sq. in.	35,000 to 40,000†	48,000 to 53,000‡
Tensile strength in gray core, lb. per sq. in.	16,000 to 25,000	22,000 to 39,000

* Symposium on Cast Iron.[582]
† Calculated from transverse tests.
‡ Test pieces ground from chilled blocks 0.800 in. in diameter, A.S.T.M. standard.

decreased almost linearly from about 70 mm. (2.756 in.) to about 10 mm. (0.394 in.) in specimens 130 × 40 mm. (5.12 × 1.41 in.) in cross-section and 150 mm. (5.91 in.) long.

E. THE HEAT TREATMENT OF GRAY IRON

Heat treatment as applied to cast iron may be divided into: (*a*) annealing at low temperatures to relieve casting strains, (*b*) annealing at higher temperatures to decrease hardness and increase machinability, and (*c*) quenching or quenching followed by tempering. The third treatment has not been used extensively; whether the strength can be appreciably increased by quenching and tempering depends upon the amount of pearlite in the as-cast material.

The heat treatment of cast iron has been competently discussed in a recent paper by Coyle.[445] A review of the literature on the heat treatment of gray iron, dealing with papers published to and including 1930, was prepared by Heller;[531] a more recent summary of work reported to 1934 was published by Bolton.[680]

122. Low-temperature Annealing.—Internal stresses resulting from casting may be removed by heating to temperatures within the range 425 to 510°C. (800 to 950°F.), holding at temperature for from 30 min. to 5 hr. (depending on the section), and cooling slowly in the furnace.[582] This treatment, which is popularly called "aging," "normalizing," or "mild annealing," has almost no influence on the mechanical properties of the iron and does not decrease the amount of combined carbon.

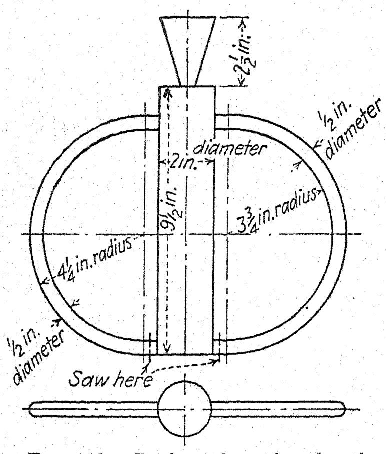

FIG. 113.—Design of casting for the measurement of the relief of internal stresses upon annealing at low temperatures. (*Walls and Hartwell.*[425])

Some data obtained by Walls and Hartwell[425] are useful in showing the influence of different annealing temperatures on the relief of internal stresses and on the hardness of the castings. Castings, of the shape shown in Fig. 113, made from a cylinder iron, were heated to different temperatures for periods of 1 and 3 hr., after which the growth was determined by measuring one dimension. One end of each of the thin parts was then severed from the vertical part and the expansion measured, the expansion as a first approximation being considered as proportional to the internal stress. The data given in Table 73 show that for annealing temperatures of 565°C. (1050°F.) or less (at which there was a considerable decrease in expansion indicating stress release) the hardness did not decrease, and that annealing at 620°C. (1150°F.) lowered the Brinell hardness by 7 points in 1 hr. and 17 points in 3 hr. Several articles by Donaldson[152,209,253] on low-temperature annealing relate both to stress release and to growth.

123. High-temperature Annealing.—Gray-iron castings may be annealed at comparatively high temperatures in order to soften the entire casting or to soften edges and corners of thin sections which are mottled and, therefore, difficult to machine. As the temperature is increased above that used for eliminating casting strains, annealing not only decreases hardness and

strength but decreases the amount of combined carbon, and with sufficient time at certain temperatures all of the combined carbon may be converted into graphite. The amount of softening and the amount of combined carbon decomposed are dependent on the composition of the iron, the softening becoming more rapid as the carbon and silicon contents increase, because these elements promote graphitization. Too long a time and too high a temperature, especially under ordinary conditions, greatly deteriorate the physical properties. Thus gray iron given a "malleable" anneal is devoid of any usable strength.

TABLE 73.—INFLUENCE OF LOW-TEMPERATURE ANNEALING TREATMENTS ON PROPERTIES OF GRAY-IRON CASTINGS*

Heating			Average growth, in.	Average expansion, in.	Brinell hardness	
Temperature		Time, hr.				
°C.	°F.				Original	Final
As cast		..	......	0.0315	179	
As cast		..	......	0.0290	187	
400	750	3	0.0002	0.0238	179	179
400	750	1	0.0006	0.0228	179	179
455	850	3	0.0012	0.0136	179	179
455	850	1	0.0012	0.0144	179	179
510	950	3	0.0010	0.0061	183	183
510	950	1	0.0017	0.0086	179	179
565	1050	3	0.005	0.0023	187	183
565	1050	1	0.0021	0.0024	179	179
620	1150	3	0.0130	0.0008	183	166
620	1150	1	0.0043	0.0004	179	172

* Walls and Hartwell.(425)

Roth(233) heated samples of cast iron containing 3.52 per cent total carbon, 0.50 per cent combined carbon, 1.75 per cent silicon, 0.75 per cent manganese, 0.05 per cent sulphur, and 0.40 per cent phosphorus to different temperatures for periods of from 30 min. to 3 hr. and cooled them in air. His results (see page 338), for heating for 3-hr. periods at different temperatures, show that the combined-carbon content decreased materially on annealing at a temperature as low as 650°C. (1200°F.), and that at 760°C. (1400°F.) almost all of the carbon was graphitized. At annealing

temperatures above 760°C. (1400°F.) the amount of combined carbon is dependent on the rate at which the material is cooled, because at the annealing temperature part of the carbon is in solution in austenite.

Annealing temperature		Combined carbon, per cent
°C.	°F.	
Unannealed		0.50
650	1200	0.15
705	1300	0.10
760	1400	0.06
815	1500	0.13
870	1600	0.21
925	1700	0.23

Figure 114 from Harper and MacPherran[87] shows the influence of different annealing temperatures on an iron containing 1.44 per cent silicon. Specimens were heated to the indicated temperatures, held at temperature for 1 hr., and then cooled in the furnace. Transverse tests were made on 1 × 2-in. rectangular bars broken on a 24-in. span. No decided change in properties occurred until the annealing temperature reached 675°C. (1250°F.). Annealing at temperatures of 675°C. (1250°F.) and above greatly decreased both strength and hardness.

124. Quenching and Tempering.—A gray cast iron which consists essentially of a ferritic matrix in which large interlocking flakes of graphite are embedded cannot be materially improved by quenching or quenching and tempering. If, however, the matrix contains considerable pearlite, or preferably if it is wholly pearlitic, and especially if the graphite flakes are small, well distributed, and do not interlock, it can be hardened by quenching from above the critical range just as steel can be hardened by quenching from the austenite field. Very few castings are quenched because: (*a*) the properties of the as-cast material are usually satisfactory for the specific application, (*b*) there is considerable danger that the castings may break when quenched into a liquid, (*c*) the strength of many gray irons cannot be appreciably increased by any quenching and tempering treatment, and (*d*) quenching and tempering is a relatively expensive

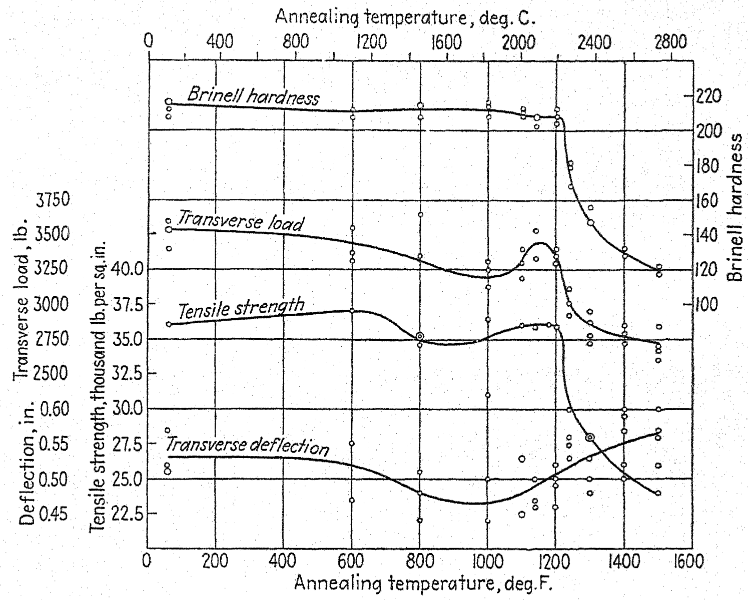

FIG. 114.—Effect of annealing temperature on properties of cast iron containing 1.44 per cent silicon. (*Harper and MacPherran.*(87))

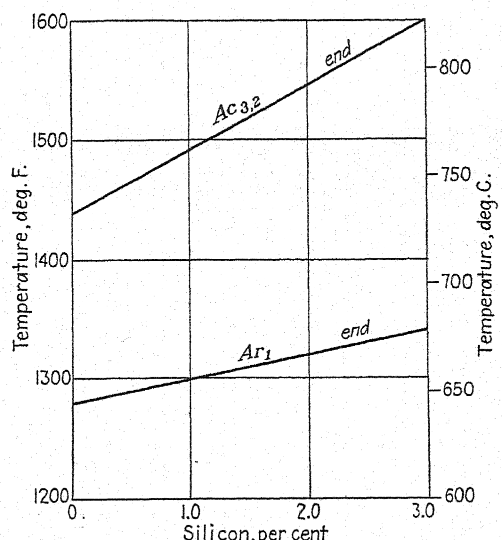

FIG. 115.—Effect of silicon on the critical ranges of cast iron containing 3.3 per cent carbon. (*Coyle.*(445))

operation. In cases where a hard (and presumably wear-resistant) material is required and low strength can be tolerated, iron is used as quenched but not tempered. In other cases, special irons are quenched and tempered in order to increase strength.

If gray iron is to be hardened by quenching, it should be quenched from above the upper critical temperature. As Fig. 115 from Coyle(445) shows, the critical temperatures of cast iron

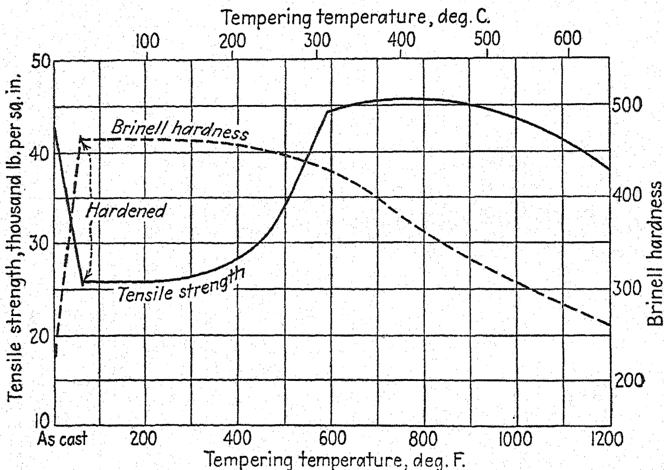

FIG. 116.—Effect of tempering temperature on the hardness and tensile strength of quenched gray cast iron. (*Symposium on Cast Iron.*(582))

increase as the silicon content increases. The figure is for iron containing between 0.55 and 0.90 per cent combined carbon. Coyle pointed out that the transformation temperatures are indistinct; those given in the figure are for the ends of the transformation ranges as shown by thermal curves.*

In determining the influence of quenching on the tensile strength of a cylinder iron, Walls and Hartwell(425) used an iron which fell within the following composition ranges:

Element	Percentage
Total carbon	3.30 to 3.35
Combined carbon	0.52 to 0.55
Silicon	2.27 to 2.33
Manganese	0.70 to 0.75
Sulphur	0.081 to 0.087
Phosphorus	0.202 to 0.210

* See "The Alloys of Iron and Silicon"(618) for complete information on transformation temperatures.

Two tensile specimens were cast in each flask, one of which was heat treated, while the other was tested as cast. Samples were quenched from different temperatures, but all of those quenched were tempered at 205°C. (400°F.). The results of the tests are given in Table 74, which indicates that by quenching from temperatures between 815 and 845°C. (1500 and 1550°F.)

TABLE 74.—INFLUENCE OF QUENCHING ON THE TENSILE STRENGTH AND HARDNESS OF GRAY CAST IRON*

Quenching temperature†		Tensile strength, lb. per sq. in.		Brinell hardness of hardened material
°C.	°F.	Original	Hardened	
650	1200	28,900	31,750	143 137
705	1300	33,700	30,500	137 140
705	1300	34,200	30,750	137 143
760	1400	34,200	31,750	137 137
760	1400	32,000	32,750	146 137
790	1450	30,100	49,500	197 341
790	1450	33,100	26,500	207 321
805	1480	32,200	28,500	388 341
805	1480	30,200	31,000	388 429
815	1500	31,600	42,500	286 444
815	1500	40,400	58,000	302 429
830	1525	36,000	40,000	444 444
830	1525	32,500	38,500	401 415
845	1550	33,300	36,000	415 415
845	1550	27,700	42,250	388 388
940	1720	28,000	19,000	444 444
940	1720	34,300	35,000	429 415
980	1800	36,000	4,500	477 477
980	1800	36,800	10,000	477 477

* Walls and Hartwell.(425)

† Bars held at the indicated temperatures for 30 min., quenched in water, and then tempered for 1 hr. at 205°C. (400°F.).

the tensile strength was increased by amounts ranging from 3000 to 18,000 lb. per sq. in. It may be well to note that there is considerable variation in the strength of specimens given identical or almost identical heat treatments.

Figure 116, from the Symposium on Cast Iron,(582) serves to indicate how the hardness and strength of quenched cast iron vary with tempering temperature.

F. AUTHOR'S SUMMARY

1. The most common method of judging the quality of gray cast iron is by the data obtained in the transverse test and the tensile test. In the former, the deflection of the center of a standard bar, tested on a definite span, and the load required to break the bar are measured. By many, the transverse test is considered as the simplest means of obtaining information on strength, flexibility, toughness, damping capacity, plastic deformation, surface condition, and machinability. Because of the deviation from Hooke's law there is no simple relation between transverse strength and tensile strength; as tensile strength increases, however, the ratio of modulus of rupture to tensile strength tends to decrease. Stress-strain curves for gray cast iron are usually curved from the origin; hence, cast iron has no true modulus of elasticity. The value for the effective modulus varies with the size and distribution of graphite from about 12 million lb. per sq. in. to as much as 22 million lb. per sq. in. for high-strength cast iron. Malleable cast iron has a modulus of about 25 million lb. per sq. in., and white iron which contains no graphite, a modulus as high as 28 million. Elastic constants are discussed in Chapter XV.

2. One of the most valuable engineering properties of cast iron is its high compressive strength which is several times its tensile strength. The ratio varies from about 3 to 1 for high-tensile cast iron to about 4 to 1 for low-strength material.

3. Another marked advantage of cast iron as a structural material is its relative insensitiveness to the effect of notches and other surface imperfections when subjected to repeated stress. This notch insensitivity and its relation to fatigue are discussed in Chapter XI. Cast iron has a low resistance to impact. Tests with the usual specimens and machines for determining impact resistance of steel are unreliable. The available data indicate that it may be possible by means of a drop test, when this is carefully standardized, to arrive at an impact value which will be a useful index of the relative amounts of plastic and elastic deformations which gray cast iron can withstand.

4. In determining the relation between composition and properties, most of the grades of cast iron may be considered as

having a high-silicon steel matrix interspersed with numerous particles of graphite. The size and distribution of the graphite particles are the most important variables in the relation between composition and mechanical properties; in general, the more finely disseminated the particles the higher the strength. If the particles of graphite are small and highly dispersed, the strength may be increased by increasing the amount of combined carbon of the matrix until this is entirely pearlitic. High-strength cast iron then consists of a matrix containing more or less pearlite and small well-distributed particles of graphite.

5. The relation between the percentages of carbon and silicon and the structure for castings of test-bar size, sand cast at normal temperature from commercial irons, has been plotted by Maurer in the form of a diagram. Later work by Maurer and associates and by other investigators has modified the diagram to show the effect of section size and of different cooling rates. These diagrams are only approximations but are of value in indicating the percentages of carbon and silicon which will, under controlled casting conditions, result in an iron with a fracture that is gray (ferritic or pearlitic), mottled, white, or a mixture of gray and mottled, or mottled and white.

6. If other variables are constant, the strength and hardness of gray cast iron increase as the percentages of carbon and silicon decrease. In most commercial cast irons, varying the amounts of manganese, sulphur, and phosphorus within the ranges usually present has little or no effect on the strength.

7. The work of a large number of investigators has shown that, in general, superheating, *i.e.*, heating molten iron considerably above the usual casting temperature, decreases the size of the graphite flakes and increases the amount of combined carbon. By controlling the carbon and silicon percentages and the temperature and time of superheating it is possible to produce high-strength cast iron by this method.

8. High-strength cast iron may also be produced by ladle additions of calcium silicide, ferrosilicon, or other silicon alloy to a low-carbon (2.0 to 3.0 per cent) iron. The composition of the original charge is adjusted, frequently by the addition of a large proportion of steel scrap, so that, as poured, the fracture of the casting would be mottled, or partly pearlitic and partly mottled, or in some cases even white. An addition of calcium

silicide, ferrosilicon, or other graphitizer, made to the molten metal before the castings are poured, results in precipitation of graphite in finely divided particles; tensile strengths as high as 55,000 or even 60,000 lb. per sq. in. are attained.

9. The present explanation for the production of finely dispersed graphite which is accompanied by high strength—which has supplanted the earlier graphite-nuclei theory—is that nonmetallic inclusions or tiny gas bubbles already present in the iron, or released as the metal freezes, are the nuclei involved in the precipitation of fine graphite particles. The mechanism of controlling the size of the graphite particles by controlling these nuclei is as readily explainable as is the control of grain size in steel by submicroscopic nuclei.

10. The mechanical properties of gray cast iron are affected by pouring temperature. The effect of this variable depends upon the composition of the iron, being more pronounced in low-carbon irons than in high-carbon materials. The size of the section has a marked effect on the mechanical properties of the casting. The available data indicate that, as the section size increases, the combined carbon is unaffected, but the size of the graphite particles increases and the strength decreases. The effect of section size is less pronounced with the stronger irons. Thus, if the tensile strength of a soft iron (25,000 lb. per sq. in.) is taken as 100 per cent for 1.2-in. diameter bars, the tensile strength of a 3-in. diameter bar will be approximately 55 per cent of this. If the tensile strength of a machinery iron (35,000 lb. per sq. in.) is taken as 100 per cent for 1.2-in. diameter bars, the tensile strength of a 3-in. diameter bar will be approximately 60 per cent of this. The exponential nature of the section sensitivity has been established by Heller and Jungbluth.

11. The composition and properties of the outer skin of cast iron differ from those of the surface after the skin has been removed. This variation in composition and in properties depends upon the mold surface and upon the way in which the particular iron is affected by different cooling rates. For most commercial applications the variation of composition and of properties of the skin is of little moment; in some cases, as for example in iron to be enameled, the variation is of prime importance.

12. Water pipe and other similar cylindrical objects are now centrifugally cast in rotating molds in large tonnages. There

are two main processes: the sand-mold process and the metal-mold process. The product of the former resembles in many of its characteristics the pipe formerly cast in stationary sand molds with the added advantage of evenness and soundness of section, improved structure and properties, and reduced cost. In the metal-mold process, the resulting pipe also has uniform wall thickness and is sound and has improved structure and properties. The surface chilling formerly characteristic of the metal-mold process is now avoided by applying a thin refractory coating to the inside wall of the metal mold. Improvement in structure and properties of gray cast iron without prohibitive decrease in machinability may also be attained by modern methods of casting in permanent metal molds.

13. The heat treatment of cast iron may be divided into: (*a*) low-temperature annealing to relieve casting strains, (*b*) high-temperature annealing to decrease hardness and improve machinability, and (*c*) quenching or quenching and tempering. Internal stresses resulting from casting are removed by heating within the range of 425 to 510°C. (800 to 950°F.), holding at temperature for 30 min. to 5 hr. (depending on the section), and cooling slowly. This treatment has practically no effect on the mechanical properties and does not decrease the amount of combined carbon. High-temperature annealing—600°C. (1110°F.) or above—is used to soften chilled edges, corners, or thin sections, or the whole casting, to improve machinability. As the temperature is increased above that necessary to relieve stresses, the hardness and strength decrease; the amount of combined carbon also decreases. The degree of softening and the amount of graphitization which take place at a definite temperature increase as the percentages of carbon and silicon increase. Too high a temperature and too long a time may cause marked deterioration of the properties. Quenching or quenching and tempering is occasionally used to increase the hardness and wear resistance. Cast irons with a ferritic matrix and coarse interlocking flakes of graphite do not harden satisfactorily; if the graphite flakes are small and well distributed and the matrix is pearlitic gray iron, castings respond to heat treatment in much the same way as carbon steel of the same carbon content. Quenching and tempering at 400 to 540°C. (750 to 1000°F.) may result in an improvement of tensile strength of 20 to 50 per cent.

CHAPTER X

MALLEABLE CAST IRON

White-heart and Black-heart Malleable Irons—Mechanical Properties of Malleable Iron—The Graphitizing Cycle for Black-heart Malleable Iron—Author's Summary

There are many industrial applications for castings for which fair strength, good ductility, and easy machinability coupled with economy of manufacture are important considerations. For the requirements in this field malleable iron* is preeminently suited. It is stronger and much more ductile than ordinary gray iron, it machines more easily than any other commercial ferrous material, and it is more readily and more economically cast into small intricate shapes than carbon steel. Because of these advantages malleable castings are widely used for agricultural implements, building equipment, electric and industrial power equipment, hardware and household appliances, machinery and machine tools, pipe fittings, and automotive and railroad equipment. A few of the many uses for malleable castings have been classified in a recent Symposium.[429] Schwartz[489] stated that the country's malleable-iron capacity is something over a million tons annually, produced by about 150 plants.

A. WHITE-HEART AND BLACK-HEART MALLEABLE IRONS

There are two steps in the production of malleable castings. In the first, iron of suitable composition is melted and poured into a mold to produce a "hard-iron" casting, one with a white fracture throughout its cross-section. In the second step, these castings, too hard and brittle to be of any value industrially,

* The term malleable iron as used in this book and in other monographs of this series refers only to castings poured as white iron and made malleable by some form of heat treatment. This footnote is made necessary by the fact that a few steel men, especially in England, still use the term to refer to wrought iron. This is another example of the unfortunate lack of uniformity in metallurgical terminology.

are subjected to a heat-treating cycle which converts the castings into a soft, ductile, easily machinable but fairly strong and tough product; the heat treatment decomposes the cementite into ferrite and free carbon—usually called "temper carbon"—which may or may not be oxidized and eliminated completely, depending upon the furnace atmosphere during the heat treatment. If the carbon is oxidized and completely eliminated, the structure consists wholly of ferrite and the fracture is white; the castings are then known as *white heart.* If the cementite is decomposed and most of the carbon remains in the casting (the carbon near the surface is usually eliminated by oxidation), the structure consists of ferrite interspersed with nodules of graphite or temper carbon, and the fracture is white at the surface but dark in the center portions; the castings are then known as *black heart.*

The practice which produces white-heart castings is used fairly widely in Europe (although it is now being supplanted by black-heart practice) but probably not at all in the United States. In 1922 Schwartz(94) stated that one American plow manufacturer was still producing white-heart malleable iron but he could cite none other in this country. Except for the brief summary of the characteristics of white-heart malleable castings given below, the discussion in this chapter is confined wholly to the production, structure, and properties of black-heart malleable iron.

125. European White-heart Malleable Castings.—In the production of white-heart malleable iron, the hard white-iron castings are annealed in a packing of iron oxide. By heating in this packing to 950 to 1000°C. (1740 to 1830°F.), holding at temperature for 60 to 80 hr., and cooling in 30 hr., decarburization of thin sections is complete. According to Schüz and Stotz,(410) irons containing 2.4 to 3.2 per cent carbon and 0.6 to 0.7 per cent silicon can be thus decarburized, provided the cross-section is less than 20 mm. (0.79 in.). Castings with carbon contents of around 3 per cent are heated to the higher temperatures of the range and for the longer times. Sections thicker than 20 mm. may not be decarburized at the center; the core, therefore, may be brittle. German specifications(410) give two classes of white-heart castings—one in which the castings should have a tensile strength of 45,500 lb. per sq. in., a yield strength of 25,000 lb. per sq. in., and an elongation of 2 per cent ($l = 5\ d$); the other in which the castings should have corresponding properties

of 54,000 lb. per sq. in., 30,000 lb. per sq. in., and 4 per cent respectively.

Schüz and Stotz indicated that slightly higher elongation values than are required by these specifications can be attained, but, at best, white-heart malleable is a brittle material. Figure 117, which shows stress-strain diagrams for European white-heart and American black-heart irons, is from their paper.

In his book on cast iron, which, incidentally, contains a detailed discussion of the production of white-heart castings, Hatfield[258]

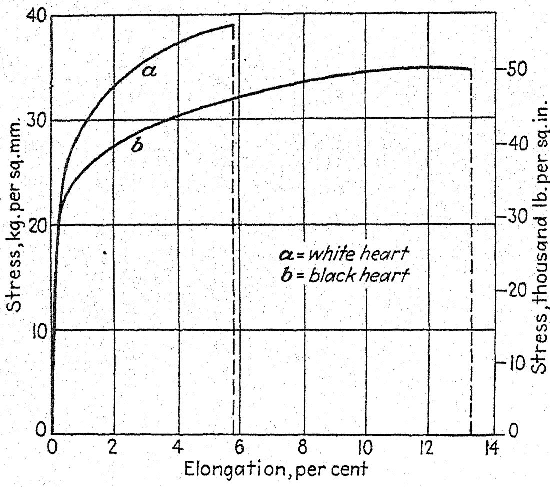

FIG. 117.—Stress-strain curves of white-heart and black-heart malleable irons. (*Schüz and Stotz.*[410])

gave average properties of European white-heart malleable and, for comparison, average properties of American black-heart. These are reproduced in Table 75. Recent data given in a Symposium[429] and by Schwartz[489] (Table 75) indicate that the properties of black-heart malleable as reported by Hatfield are much lower than they are now regularly obtained in this country, especially the values for ductility as indicated by the percentage elongation.

126. Production of Black-heart Malleable Castings.—White-iron castings to be malleablized by the black-heart process are poured from a charge of malleable pig iron (see page 262, for composition), white-iron sprues from previous heats, malleable-iron scrap, and steel scrap. In a large majority of plants the

metal is melted in air furnaces, usually two heats of 8 to 24 tons each per day. Cupola melting is also used but is not recommended for the production of high-grade castings. Small open-hearth furnaces of 10 to 20 tons capacity, and electric furnaces of 5 to 8 tons capacity are also used to a limited extent. The melting practice for the production of castings to be malleablized has been described in so many elementary textbooks on metallurgy that further discussion here is unnecessary. A full description, including extended comments on the various furnaces used and the advantages and disadvantages of each type, is given by Schwartz[94] in his book on malleable cast iron.

TABLE 75.—AVERAGE MECHANICAL PROPERTIES OF WHITE-HEART AND BLACK-HEART MALLEABLE CASTINGS

Property	Hatfield[258]		Symposium[429]	Schwartz[489]
	European white heart	American black heart	American black heart*	American black heart†
Tensile strength, lb. per sq. in.	42,600 to 65,000	35,800 to 58,200	50,000 to 59,000	54,000
Yield strength, lb. per sq. in.	26,800 to 40,300	22,400 to 44,800	34,000 to 40,000	36,000
Elongation in 2 in., per cent.	2 to 6	4.5 to 15	10 to 30	18
Reduction of area, per cent.	2 to 6	4.5 to 15		

* Values in this column are the properties determined in approximately 99 per cent of 5000 tests.

† Average properties of 20,000 heats.

The composition of the charge from which white-iron castings are poured varies widely according to the melting process used, the size of the castings, and the properties desired after malleablizing. For air-furnace practice, the average composition, according to a Symposium,[429] and the average losses in melting, according to Schwartz,[94] are as shown in the table on page 350.

In determining the composition of the white-iron castings and of the furnace charge from which they are poured, several conflicting factors must be considered. First and most important: the iron must be graphite free after solidification in the mold,

but for economical production the iron carbide must dissociate into ferrite and nodular graphite when the castings are heated to a reasonable temperature for a reasonable time. The lower the temper carbon the greater are the strength and ductility (contrary to carbon steel the percentage elongation in malleable castings increases as the tensile strength increases), but decreasing the carbon increases the foundry difficulties.

Element	Composition of charge, per cent	Loss in melting, per cent
Carbon	2.50 to 3.00	15.8
Silicon	1.20 to 1.30	31.4
Manganese	0.55 to 0.65	48.1
Phosphorus	0.18 or less	None
Sulphur	0.065 or less	22.2 (gain)

Schwartz[489] summed up the question of composition by stating that the carbon may run from 2.0 to 2.2 per cent where strength and ductility are major considerations, from 2.3 to 2.6 per cent for a fair combination of all desired properties, and from 2.8 to 3.0 per cent where prevention of foundry difficulties is a major item. The low-carbon irons will contain around 1 per cent silicon, the intermediate 0.8 to 0.9 per cent, and the higher carbon irons around 0.7 per cent.

Since the final ferrite is practically carbonless, it can hold as much as 0.2 per cent phosphorus without becoming embrittled (after the normal anneal and no further heating); and since malleable iron is not hot worked, sulphur can range from 0.06 to 0.15 per cent for iron melted in an air furnace, and from 0.15 to 0.25 per cent in cupola malleable. Manganese should be present in amounts not much more than sufficient to combine with the sulphur to form manganese sulphide, hence, it is generally under 0.3 per cent in iron which has been melted in an air furnace, and from 0.4 to 0.6 per cent in cupola iron. Manganese tends to stabilize cementite and may thus retard the malleablizing process; hence, any excess above the amount necessary to combine with sulphur is kept as small as possible. On the other hand, the formation of iron sulphide must be avoided as this constituent not only also retards the dissociation of cementite but in addition

may form harmful intergranular films. Regarding sulphur, Schwartz[489] stated: "Curiously enough, it has been found that it is actually disadvantageous to make metal very low in sulphur on account of the difficulties with regard to the formation of graphite during freezing."

As has been indicated before, the composition must be adjusted according to the size of the casting, in other words according to

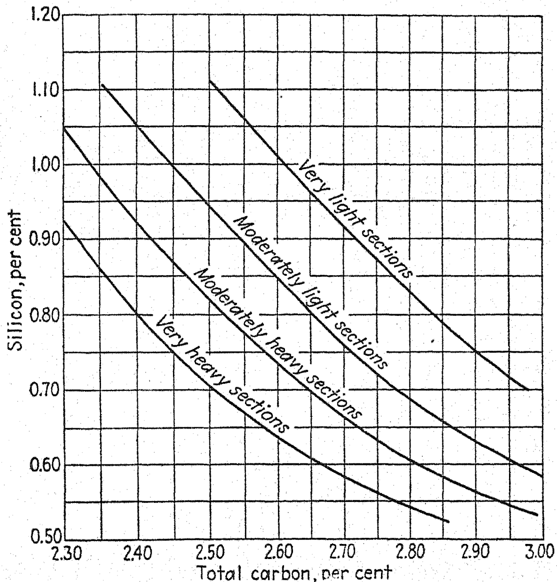

FIG. 118.—Relation between the silicon and carbon contents of malleable iron and the size of section. (*Schwartz*[94] *after Thrasher.*)

the rate at which the metal solidifies. The relation between carbon and silicon percentages and the size of the casting is shown in Fig. 118. Although Schwartz did not explain what is meant by heavy and light sections, it may be assumed that very heavy sections are 1.25 in. thick or over and that very light sections are ⅜ in. or less. To secure sound castings, the minimum silicon content should be about 0.60 per cent. Since this element promotes dissociation of cementite, the higher the percentage, provided of course that premature graphitization does not take place in freezing, the more readily the casting graphitizes when annealed.

Silicon is also of benefit for its strengthening effect. Ferrite which contains this element in solution, within the limits met with in malleable iron, has higher tensile and yield strengths but no lower ductility than pure alpha iron; hence, increasing the silicon as much as possible (provided, of course, that not enough is present to cause premature graphitization) results in a malleable iron with the highest strength for a given carbon content.* According to curves given by Yensen,[37] ferrite containing 0.5 per cent silicon has a tensile strength of 42,000 lb. per sq. in. and a yield strength of 24,000 lb. per sq. in. Increasing the silicon to 1.1 per cent increases these values to about 47,000 and 31,000 respectively. In annealed silicon ferrite, free from phosphorus and temper-carbon particles, the elongation for materials of 0.5 and 1.1 per cent silicon is about 50 per cent and the reduction of area about 90 per cent.

127. Miscellaneous Physical Properties of Black-heart Malleable.—All white cast iron of the composition suitable for malleablizing shrinks linearly from the pattern size by practically the same amount on freezing;[429] but, since cementite occupies a smaller volume than its decomposition products, iron and temper carbon, the castings expand on annealing to varying degrees depending on the amount of carbon. A carbon-content versus contraction curve published in the malleable-iron Symposium[429] shows that the shrinkage allowance from the pattern size for a carbon content of about 3.06 per cent is 1 per cent, and for a carbon content of about 2.47 is 1.5 per cent. Between these two points the curve is linear. Specific gravity also varies with the carbon content[429] from 7.15 to 7.45 at 20°C.

For thermal expansion the Symposium[429] reported a value of 0.000012 per °C. (0.0000066 per °F.) for the coefficient as sufficiently accurate in ordinary computations for temperatures up to 400°C. (750°F.). The same authority gave the thermal conductivity (in cal. per sec. per sq. cm. per °C. per cm.) as 0.145 at 50°C., 0.137 at 100°C., and 0.115 at 200°C., stating that "the data can be taken only as representing a first approximation at the true values, but may be useful lacking better knowledge."

Direct calorimetric determinations of mean specific heat of malleable iron between room temperature and various higher temperatures are as follows:[429]

* "The Alloys of Iron and Silicon,"[618] pp. 78–85.

Temperature Range, °C.	cal. per g. per °C.
20 to 100	0.122
20 to 200	0.125
20 to 300	0.128
20 to 400	0.133
20 to 500	0.139
20 to 600	0.146
20 to 700	0.159

The specific heat at room temperature, calculated from the specific heat of iron and graphite, is 0.1102 cal. per g. per °C.

The electric resistivity of malleable iron varies with the carbon content from about 28 to 37 microhms per cu. cm. An average of 37 tests(429) gave a value of 30.5. According to Schwartz,(94) the resistivity increases with the temperature, but the increase is not linear. Resistivity is doubled at approximately 425°C. (795°F.) and tripled at 635°C. (1175°F.). Magnetic properties for various malleable irons are given by the Symposium,(429) but as these are not of unusual interest the curves are not reproduced here. (Further information on the foregoing properties is given in Chapters XV and XVI.)

128. Machining and Welding.—It has been mentioned above that one of the outstanding advantages, one which is in a measure responsible for the wide use of malleable iron, is its excellent machinability. Except for the data by Boston,(439) there is little information available. In fact, according to the Symposium,(429) "the machinability of malleable iron cannot be dealt with quantitatively since there is no generally accepted test procedure or unit of measurement." It is, however, generally accepted that "as measured by tool life or cutting speed in commercial operations, malleable iron machines more readily than any other ferrous material." Opinions differ whether variations in tensile strength are accompanied by variations in machinability, but "it may be said that for practical purposes such variations are not sufficient to contradict the preceding statement for any normal grade of malleable iron." This assumes, of course, that no pearlite is present.

The data by Boston(439) are plotted in Fig. 119, in which the materials studied are arranged in the order of ease of planing. The cast iron, in the form of 1-in. sq. bars 12 in. long, and the malleable iron, in 1⅛-in. sq. sections, had the following compositions:

Element	Percentage	
	Gray cast iron	Malleable iron
Total carbon	3.53	2.19
Combined carbon	0.53	
Graphite	3.00	2.17
Silicon	....	0.87
Manganese	....	0.26
Phosphorus	....	0.15
Sulphur	....	0.09

It will be noted from Fig. 119 that the materials do not necessarily arrange themselves in the same order for other types of

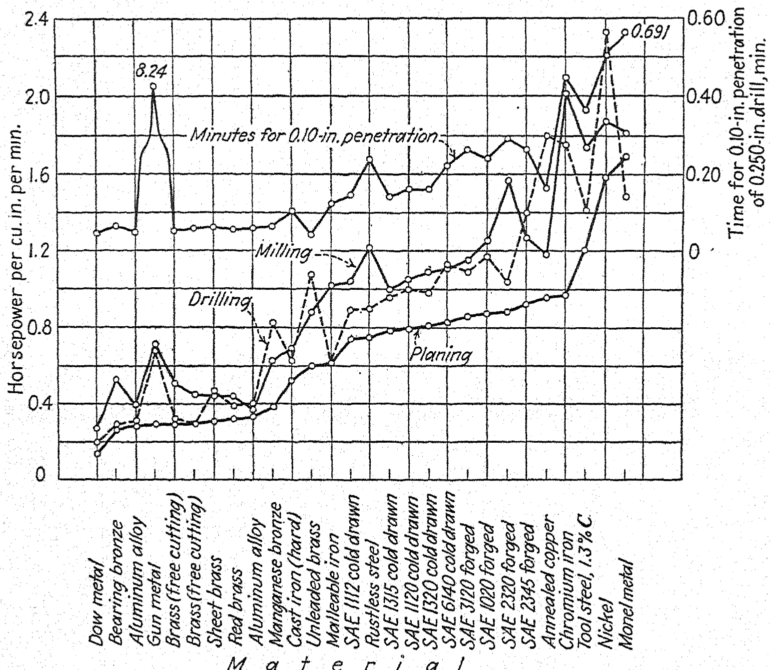

FIG. 119.—Relative machinability of malleable iron as compared with other materials. Low value for horse-power indicates easy machinability. (*Boston.*[439])

machining. Schwartz[411] concluded that, within the commercial range of carbon contents of normal malleable iron, the higher

the carbon content of the casting the more readily it is drilled.

Since reheating malleable iron above the lower critical point dissolves some of the temper carbon, welding is not commercially feasible. An annealing treatment comparable to the regular malleablizing treatment would be necessary to restore the original ductility.

B. MECHANICAL PROPERTIES OF MALLEABLE IRON

A large proportion of the malleable iron used in the United States is purchased under specifications which give minimum values for tensile strength, yield strength, and elongation. Most of the current specifications are issued by the railroads which are large users of malleable castings; all of them are listed in Standards and Specifications for Metals and Metal Products issued by the National Bureau of Standards.* Most of the specifications for air-furnace iron conform to No. A47-30, issued in 1930 by the American Society for Testing Materials,[429] and for cupola iron to No. SP-22-1931, issued in 1931 by the Manufacturers' Standardization Society of the Valve and Fittings Industry.[429] The requirements of these specifications are:

Specification		Minimum properties*		
Number	For	Tensile strength, lb. per sq. in.	Yield strength, lb. per sq. in.	Elongation in 2 in., per cent
A47-30	Air-furnace malleable......	50,000	32,500	10
SP-22-1931	Cupola malleable..........	40,000	30,000	5

* Reduction of area is not specified and is seldom determined. It is normally around 20 per cent. Malleable iron does not neck down appreciably but stretches throughout the gage length.

129. Mechanical Properties of Cupola Malleable.—White iron which has been melted in a cupola has some advantages[429] over air-furnace iron which make it especially suited for such castings as valve and pipe fittings. Among these advantages may be mentioned lower shrinkage from the pattern size and greater

* *Misc. Publ.* 120, 1933.

fluidity (because the carbon is generally higher than in air-furnace iron), machinability even superior to that of air-furnace malleable, absence of brittleness in galvanizing, and tightness and strength under water or steam pressure. As is shown below,

TABLE 76.—TENSILE PROPERTIES AND RESULTS OF BURSTING TESTS OF CUPOLA-MALLEABLE IRON*

Tensile properties

Property	Number 1	Number 2	Number 3
Tensile strength, lb. per sq. in.			
Maximum	58,200	52,000	45,000
Minimum	41,300	35,000	40,000
Average†	49,700	43,000	43,000
Yield strength, lb. per sq. in.			
Maximum	46,800	37,000	34,000
Minimum	34,300	27,000	30,000
Average†	41,000	33,000	31,000
Elongation in 2 in., per cent			
Maximum	16.0	12.0	8.0
Minimum	5.8	2.0	5.0
Average†	8.1	7.0	6.5

Water-bursting tests

Size of fitting, in.	Ells, average maximum pressure, lb. per sq. in.		Tees, average maximum pressure, lb. per sq. in.	
	Cupola	Air furnace	Cupola	Air furnace
0.25	8750	7086	7600	4600
0.75	8833	8285	6000	7071
1.25	7000	4714	5250	4385
2	4616	4528	3733	3357
3	....	....	3067	4333
6	....	....	2550	2500

* Symposium.(429)

† In a recent publication (American Foundrymen's Association, *Preprint* 35-19, 1935) F. B. Riggan reported average properties of several hundred bars (5/8-in. diameter) of cupola malleable as 48,000 lb. per sq. in. tensile strength, 34,000 lb. per sq. in. yield strength, and 9 per cent elongation in 2 in. Machining 1/8 in. from the surface of the test bars had no appreciable effect on the properties.

the actual bursting strength of cupola malleable—an important service requirement for pipe fittings—is generally higher than that of air-furnace castings.

Tensile properties of cupola malleable as supplied by three representative manufacturers and published in the Symposium[429] are given in Table 76 together with the results of water-bursting tests of cupola and air-furnace pipe fittings. For

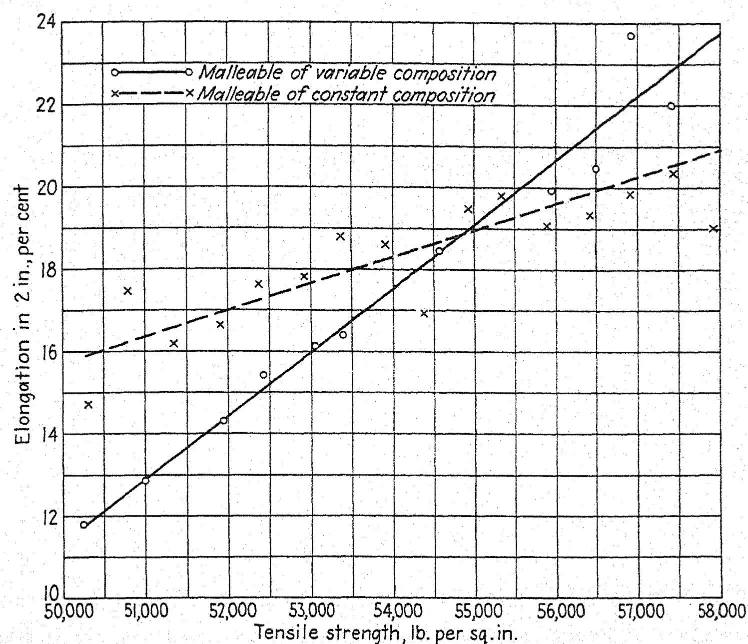

FIG. 120.—Relation of tensile strength to elongation in air-furnace malleable iron. (*Symposium on Malleable Iron Castings.*[429])

the latter tests commercial fittings, all made for the same types of service, were purchased. The pressures reported are the maxima in lb. per sq. in., obtained before the fitting burst or leaked too badly to permit higher pressures. Values reported for the four smaller sizes are averages of six tests on cupola fittings and seven tests on air-furnace fittings. For the 3- and 6-in. fittings the values are the averages obtained on fittings from four and three manufacturers respectively.

If the tensile values given in Table 76 are compared with the values given in Fig. 120 for air-furnace iron, it will be noted that

the cupola iron is somewhat lower in strength and much lower in ductility than the product of the air-furnace. In resistance to bursting, however, cupola iron is apparently superior to air-furnace material; this is especially pronounced in the fittings of smaller size.

The next section (page 362) includes a brief résumé of German black-heart malleable practice. The tensile properties obtained by commercial cupola operation in Germany are apparently somewhat better than the best combination of properties shown in Table 76.

130. Mechanical Properties of Air-furnace Malleable.—In preparing material for the 1931 Symposium on Malleable Iron Castings[429] Toúceda analyzed the results of 5000 tensile tests on air-furnace malleable made over a period of 6 months by 52 companies. These tests are representative of so-called "standard specification" malleable (see specification A47-30, page 355). The results are presented in the Symposium in tables and in frequency curves. If the values are classed as low, normal, and high, according to the following:

Property	Low	Normal	High
Tensile strength, lb. per sq. in.	Less than 50,000	50,000 to 59,000	Over 59,000
Yield strength, lb. per sq. in.	Less than 34,000	34,000 to 40,000	Over 40,000
Elongation in 2 in., per cent	Less than 10	10 to 30	Over 30

it will be noted that approximately 99 per cent of the malleable iron made commercially falls within the class selected as normal, *viz.*:

Property	Average of 5000 tests	Percentage of whole number of tests which may be classed as		
		Low	Normal	High
Tensile strength, lb. per sq. in.	54,000	0.74	98.70	0.56
Yield strength, lb. per sq. in.	36,600	0.04	99.74	0.22
Elongation in 2 in., per cent	19.3	0.92	98.90	0.18

It has been mentioned that malleable iron is unique in that the percentage elongation increases with the tensile strength. The relation between the two has been investigated and the results have been reported by the Symposium.[429] The 11 plants which participated divided their production (40 to 1300 heats for each plant) into groups of approximately constant chemical composition and averaged the tensile strength and elongation of each group. In Fig. 120 the mean tensile properties (open circles) are plotted by groups for all plants. In addition, each plant studied the properties of specimens of a single group of nearly constant composition, the group being that which, considering all of the plants, represented the greatest number of specimens. These data are shown by the crosses in Fig. 120. The curves in Fig. 120 were drawn by inspection; this was considered to be sufficiently accurate. In commenting upon the results it was stated:

> . . . insofar as tensile strength is determined by original composition, an increase of 5000 lb. per sq. in. in tensile strength is accompanied by an increase of about 8 per cent in elongation [solid curve in Fig. 120]. If the variation in tensile strength is due to factors other than composition, an increase of 5000 lb. per sq. in. in tensile strength is accompanied by an increase of only about 3 per cent in elongation [dashed curve in Fig. 120]. No definite reason for this difference can be assigned. The mechanical properties of malleable iron are determined mainly by carbon content and this in turn by both original content and by decarburization [in annealing]. It is suggested that decarburization affects tensile strength more than it does elongation, possibly through the intervention of oxygen or small amounts of combined carbon.

Although it may be stated that, in general, surface decarburization affects the mechanical properties favorably, removing this decarburized zone by machining has little or no effect on the strength and not much effect on elongation.* Data were presented in the Symposium which show that machining ¾-, ⅞-, and 1-in. cast bars to ⅝-in. diameter has no appreciable influence on tensile and yield strengths, as compared with the standard ⅝-in. cast bar. In four out of six sets of tests, removing the surface lowered the elongation to 80 or 90 per cent of the value on the as-cast bar; in the other tests machining lowered it very

* Riggan has shown this to be the case also for cupola malleable. See footnote to Table 76.

little. Further tests indicated that, as compared with the properties of cast test bars, machined bars in some cases had slightly lower properties, in other cases slightly higher properties, and in still other cases practically the same properties.

The effect of cross-section on the properties of malleable-iron bars as cast to size and as cast and machined to remove 1/16 in.

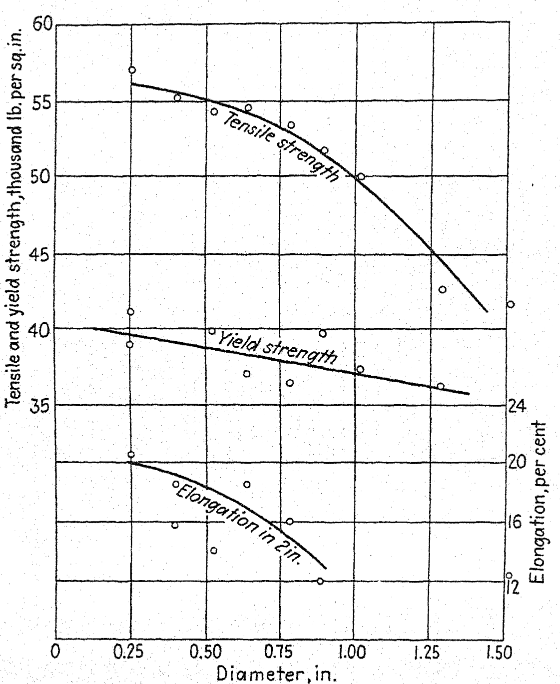

FIG. 121.—Effect of increasing section on the tensile properties of malleable iron. Specimens not machined. (*Symposium on Malleable Iron Castings.*[429])

of surface metal was investigated by Schwartz.[63] The iron as cast into the test specimens contained 2.44 per cent carbon, 0.76 per cent silicon, 0.18 per cent manganese, 0.166 per cent phosphorus, and 0.099 per cent sulphur. Conditions were carefully controlled so that no primary graphite would form in casting, so that the surface would be as smooth as possible, and so that the bars would be free from shrinks and other internal defects which might affect the properties. The variation of properties with the cross-section of the specimen is shown for the as-cast specimens in Fig. 121 and for the machined specimens

in Fig. 122. Considering the results of Schwartz and the other data summarized on the preceding pages, the Symposium

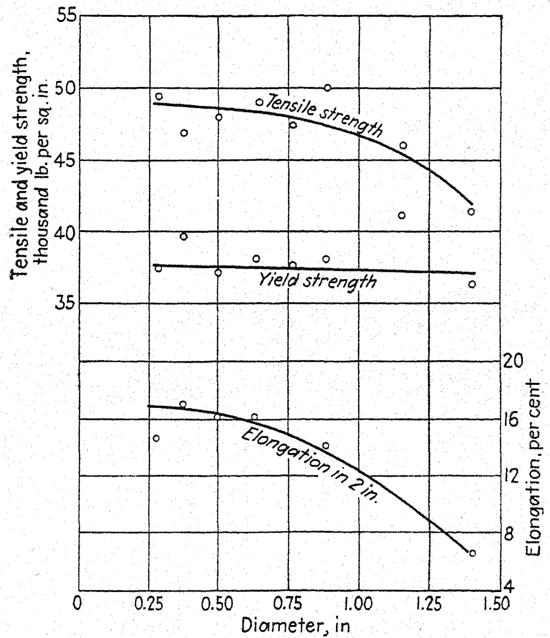

FIG. 122.—Effect of increasing section on the tensile properties of malleable iron. Machined specimens. (*Symposium on Malleable Iron Castings.*[429])

presents the following general conclusions regarding the effect of decarburization and of cross-section on the mechanical properties:

1. Decarburization has a favorable influence upon the strength and ductility of the product.
2. The effect of quick cooling in freezing on the surface metal of a casting is such as to improve the strength and ductility of the product.
3. Roughness of surface of a cast specimen apparently decreases the strength and especially the ductility.
4. The tensile strength decreases with increasing diameter of section by an amount proportional to the cube of the diameter.
5. The elongation decreases with increasing diameter of section by an amount proportional to the $5/2$ power of the diameter.
6. The yield point is apparently not affected by any of the variables investigated.

In addition to the variables already discussed, the elongation of malleable iron may be affected by such factors as the method of

gating, which controls the feeding of the casting and its freedom from internal shrinkage, and by the completeness of the anneal, since any appreciable residue of untransformed cementite lowers the ductility. The more uniform the distribution of the temper-carbon nodules throughout the mass, the better is the ductility. There are various factors not readily defined, which may influence this distribution and thus affect the ductility. Experience, rather than adherence to a set of readily definable rules, seems to be important in securing ductility.

Some of the other mechanical properties of black-heart malleable iron may be summarized briefly as follows: (1) the modulus of elasticity in tension or compression is 25,000,000 lb. per sq. in., (2) Brinell hardness ranges from 100 to 140, averaging about 115, (3) in compression a permanent set of 1 per cent is attained at about 28,000 lb. per sq. in., while the yield strength in torsion is about 25,000 lb. per sq. in. and in shear about 23,000 lb. per sq. in.; and (4) the ultimate shearing strength is about $\frac{9}{10}$ of the tensile strength, and the modulus of rupture in torsion is 58,000 lb. per sq. in.

Schüz and Stotz[(410)] furnished some interesting data on German practice in making the American-type black-heart malleable, though details of composition and annealing cycle were not stated. In one plant, cupola black heart averaged over a week's period 54,000 lb. per sq. in. tensile strength (range 50,000 to 60,000) and 11 per cent elongation (range 9 to 14 per cent) on a 60-mm. (2.36-in.) test length, obtained by careful control of casting and annealing. A plot of a larger number of heats showed variations in tensile strength from 50,000 to 72,500 lb. per sq. in. and 7 to 14 per cent elongation.

Similar data for an open-hearth malleable gave 50,000 to 70,000 lb. per sq. in. tensile strength (average 57,000) and 9 to 14 per cent elongation (average 11); for malleable iron from an oil-fired furnace values ranged from 50,000 to 58,000 (average 52,000) and from 9 to 16 per cent (average 12), while those for malleable from a coal-fired Brackelsberg furnace were 57,000 to 68,000 (average 63,500) and 14 to 22 per cent (average 19). Roesch[(758)] also discussed German melting practice.

The Sesciand-Brackelsberg furnace and other furnaces of similar type are generally considered to be more readily controlled at high melting temperatures than the other types of fuel-

fired furnaces. According to Schwartz,* American practice indicates that iron melted in the Brackelsberg furnace does not differ in properties from similar iron melted in an air furnace.

131. Effect of Superheating Malleable Iron.—Superheating the melt has been suggested[671] as being helpful in the production of malleable iron and in reducing the annealing time. Tanimura,[495] whose work on cast iron has been mentioned on page 312, reported data on the malleablization of superheated irons. They were made from white pig iron, charcoal, and ferrosilicon and contained: (1) 3.62 per cent carbon and 0.30 per cent silicon, (2) 3.60 per cent carbon and 1.37 per cent silicon, and (3) 3.72 per cent carbon and 2.07 per cent silicon. They were held for 10 min. at temperatures from 1400 to 1600°C. (2550 to 2910°F.), cooled to 1350°C. (2460°F.), poured into chill molds, and then malleablized. He reported that graphitization did not start so quickly in those irons heated to the lower part of this range as in those held at 1600°C. (2910°F.), but when graphitization did start it proceeded more rapidly. Tanimura assumed that, even though the iron is fully white, tiny graphite nuclei are present after the lower melting temperatures, while at the higher ones they are absent which permits strong undercooling to occur. In reply to discussion of one of his papers Piwowarsky[169] stated in 1925 that with a melting temperature of 1600°C. (2910°F.) the temper carbon in malleable iron separated in a finer form.

White and Schneidewind,[671] in an investigation made primarily to determine if the annealing time could be reduced by superheating, determined the properties of malleable irons after superheating to various temperatures. Compositions, details of superheating, malleablizing time, and the resulting properties are given in Table 77. Iron 1 was cupola melted under regular production conditions and was superheated in an electric-arc furnace. The other two irons were special low-carbon alloys, melted and superheated in a rocking electric furnace.

From the data shown in Table 77 these investigators deduced that it would be possible to choose a composition and superheating practice which would result in a strong and ductile iron and which, with the use of a tunnel kiln, would allow a reduction in annealing time from the normal 145 hr. to around half that time.

* Private communication.

TABLE 77.—EFFECT OF SUPERHEATING ON THE PROPERTIES OF MALLEABLE IRON*

Specimen number	Composition, per cent					Superheating			Malleablizing time, hr.	Properties		
	C	Si	Mn	P	S	Furnace	Temperature °C.	Temperature °F.		Tensile strength, lb. per sq. in.	Yield strength, lb. per sq. in.	Elongation in 2 in., per cent
1*A*	2.80	1.19	0.36	0.08	0.095	Electric arc	1630	2970	51.5	49,500	46,300	14.0
1*B*	2.71	1.17	0.35	0.08	0.095		1595	2900	61.0	49,100	46,000	12.5
1*C*	2.71	1.18	0.35	0.08	0.095		1510	2750	70.5	49,000	40,000	14.1
2*A*	1.4	1.5	....	....	.....	Rocking electric	1730	3150	12.0	64,800	47,400	12.3
2*B*							1575	2865	20.0	61,700	58,000	18.0
3*A*	1.5	1.7	....	....	.....	Rocking electric	1750	3180	5	63,700	49,200	10.4
3*B*							1665	3030	8.75	62,200	42,400	18.5
3*C*							1565	2850	11.5	Bars unsound		

* White and Schneidewind.(671)

132. Improving the Mechanical Properties of Malleable Iron.—In the past 15 to 20 years much attention has been paid to improving the properties of black-heart malleable. The chief means of effecting this improvement has been to lower the carbon content of the original charge from about 2.35 to 2.60 per cent for "standard malleable" to about 2.00 to 2.20 for "high-strength malleable." It has been necessary to accompany this lowering of the carbon by careful supervision of melting, including close control of the composition, precautions to insure soundness, and control of the other variables in the entire process of manufacture.

The production of high-strength malleable has in some measure been due to the demands for such a product on the part of the railroads. The Malleable Iron Research Institute has recently adopted specifications[429] for malleable iron for railroad work, with a minimum yield strength of 35,000 lb. per sq. in. and a minimum elongation of 18 per cent, and has recommended that the American Society for Testing Materials and the American Railway Association adopt similar specifications.

TABLE 78.—AVERAGE MECHANICAL PROPERTIES OF BLACK-HEART MALLEABLE IRON 1916 TO 1931

Authority	Year	Source of data	Tensile strength, lb. per sq. in.	Yield strength, lb. per sq. in.	Elongation in 2 in., per cent
Schwartz[94].......	1916	Average properties, specimens tested monthly by American Malleable Castings Association	49,000	......	10
Schwartz[94].......	1921	Same as above	53,000	......	16
Symposium[429]....	1931	Average of 5000 tests (see p. 358)	54,000	36,600	19.3
Symposium[429]....	1931	High-strength malleable, 236 tests, 4 foundries, average maximum	61,900	39,900	28.2
		Same, average minimum	53,100	35,800	19.4
		Same, grand average	57,600	37,900	23.1

The average properties of black-heart malleable in 1916 and 1921 and the grand average of 5000 tests reported in the symposium in 1931 are given in Table 78. Included also in this table are the average minimum, average maximum, and the grand average of 236 tests of high-strength malleable made by four foundries

under a specification which required a yield strength of 35,000 lb. per sq. in. or more and 18 per cent or more elongation.

As early as 1922 Schwartz[94] recorded a specimen with a tensile strength of 58,000 lb. per sq. in. and 34 per cent elongation; he commented that any malleable iron with a strength approaching 60,000 lb. per sq. in., without at least 12 per cent elongation, would be considered to be the result of poorly controlled practice.

In the discussion of the symposium,[429] Wolf and Meisse pointed out that, by the statistics presented, a third of the malleable iron regularly produced in the United States would not meet the minimum requirement of 18 per cent elongation as suggested for railroad work. The authors of the symposium replied that the specimens tabulated were produced under a specification with a requirement of 10 per cent or more elongation and that, if the minimum requirements were raised to 18 per cent, producers would be forced to exercise special control of their manufacturing processes to meet this requirement.

Recently, considerable interest has been shown in producing high-strength malleable by a special short-cycle anneal of an iron of suitable composition or by alloying. In 1933, Schwartz, in discussion of a paper by Hall,[620] pointed out that a considerable range of commercial materials, differing from malleable by having a pearlitic matrix instead of a ferritic one, is now obtained either by modifying the anneal so that some pearlite remains, or by reheating normal malleable to reform pearlite or sorbite, or by introducing alloying elements which retard the decomposition of cementite to temper carbon during the normal annealing cycle.

C. THE GRAPHITIZING CYCLE FOR BLACK-HEART MALLEABLE IRON

The process of graphitizing white-iron castings consists, as has been stated, of packing the castings in pots, slowly heating to a temperature of 845 to 925°C. (1550 to 1700°F.), holding at this temperature for a time sufficient for graphitization to become complete, and cooling again. Although graphitization is a process which does not involve any change in chemical composition (except some surface decarburization) and will take place in the absence of any packing material, it is customary to pack the castings in scale or slag which consists essentially of

ferrous silicate. Raw packings, such as iron ore or unused scale, which contain a large amount of free iron oxide are avoided; the addition of such material is limited to a small amount for each charge. This limits the amount of ferrous oxide (FeO) available for reaction with iron carbide and thus restricts the surface decarburization.

The graphitization cycle has been discussed in detail by Storey,[30] Phillips and Davenport,[92] Hayes and coworkers,[157,158]

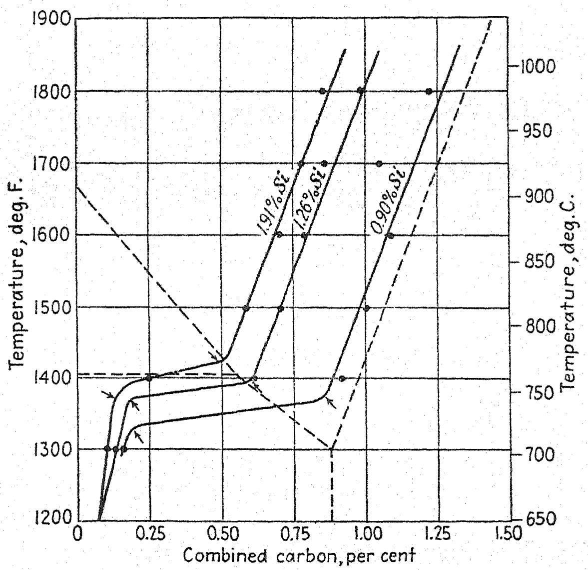

FIG. 123.—Effect of silicon on the location of the A_{cm} and A_1 lines in the iron-carbon diagram. (*White and Schneidewind.*[579])

Kikuta,[184] Schwartz,[197] Schwartz and Johnson,[198] and by White and Schneidewind.[579]

133. The First Stage of the Cycle.—The binary iron-carbon equilibrium diagram is not directly applicable to the graphitizing process because of the presence of silicon in malleable iron. The effect of silicon on the solubility of carbon in austenite, as determined by White and Schneidewind, is shown in Fig. 123; these solubility lines for several constant silicon contents are parallel to the A_{cm} line.

The first stage of graphitization is accomplished by slowly heating (for about 2 days) the charge into the austenite field—say 845 to 925°C. (1550 to 1700°F.) but usually to about 870°C.

(1600°F.)—and holding it there from 48 to 60 hr. At the beginning of this soaking period the material consists of saturated austenite and free cementite. As Fig. 123 shows, in a 0.9 per cent silicon malleable iron at 870°C. (1600°F.) about 1.1 per cent carbon is in solution in austenite, the remainder existing as free cementite, which is not stable and which slowly transforms into fine graphite (temper carbon). At extremely high temperatures it tends to decompose into flakes somewhat similar to those in gray iron. This must be avoided.

When all excess cementite has been transformed to iron and graphite, the "first stage" is ended. The transformation progresses in the way characteristic of many chemical reactions, *i.e.*, it begins slowly (this is known as the period of induction), then proceeds rapidly in a way such that the reaction rate at any instant is proportional to the amount of untransformed cementite. (This behavior is illustrated in Volume I of this monograph on page 186.) This behavior implies that, at a constant temperature of transformation, the logarithm of fractional cementite decomposition is a linear function of time. Further to be noted is the fact that the reaction approaches its end slowly.

From the form of the temperature coefficient of such reactions it can be deduced that transformation time is a logarithmic function of reciprocal absolute temperature. Thus, according to the principles of the kinetics of transformation, Fig. 124, from White and Schneidewind,(579) is incorrectly plotted.

Silicon affects the value of the velocity constant of the transformation; this shifts the lines of log time versus reciprocal absolute temperature in a manner analogous to the shift observable in Fig. 124. Increased silicon decreases the time required to complete the first stage.

134. The Second Stage of the Cycle.—The next step, the second stage, involves precipitation of the carbon held in solution in austenite, and the graphitization of the cementite thus formed. This is accomplished by slowly lowering [about 5°C. (10°F.) per hr.] the temperature; more cementite separates from the austenite as the material cools to the temperature of the eutectoid. Above the eutectoid temperature, as the temperature drops, cementite is slowly precipitated and, if the temperature fall is slow enough, graphitization can keep pace with it. As the temperature drops

below the eutectoid, all the remaining austenite changes to ferrite and cementite; the remaining cementite has, therefore, still to be graphitized. This is accomplished by continuing to cool slowly, or by holding just below the critical range.

As the silicon percentage increases, the carbon content of the eutectoid decreases; hence, the higher the silicon the less cementite is present to be graphitized below the critical range, provided, of course, that the cooling has been slow down to that

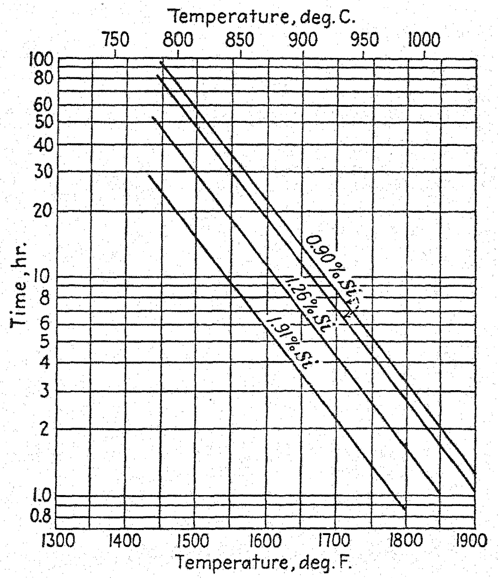

FIG. 124.—Effect of temperature on the time for first-stage graphitization. (*White and Schneidewind.*[579])

point. It is possible to carry out the first stage by holding in the austenite range, say at 870 or 925°C. (1600 or 1700°F.), then to drop the temperature fairly rapidly to just below the critical and hold the material at this temperature until the second stage is completed. This can be done in tunnel-kiln types of annealing furnaces, but the annealing pots and packings of common graphitizing practice have so much heat capacity that slow cooling is more convenient.

White and Schneidewind[579] have worked out the minimum graphitizing cycles for air-furnace malleable containing 2 to 2.5 per cent carbon. Their data are shown in Table 79. The

total given in the last column of the table does not include the time of heating the charge to the first-stage temperature—fixed by White and Schneidewind at 925°C. (1700°F.)—nor the time of cooling from the second-stage temperature, 720°C. (1325°F.), to approximately room temperature.

135. Effect of Pre-quenching.—Saito and Sawamura(234) and Sawamura(338) reported that white iron which had been quenched from 850 to 1000°C. (1560 to 1830°F.) graphitized more readily

TABLE 79.—MINIMUM MALLEABLIZING CYCLES FOR AIR-FURNACE IRONS OF 2 TO 2.5 PER CENT CARBON*

Silicon, per cent	Time at 925°C. (1700°F.), hr.	Time from 925°C. (1700°F.) to A_1, hr.	Time at 720°C. (1325°F.), hr.	Total time, hr.
0.7	20.0	20.0	85	125.0
0.8	16.0	16.0	65	97.0
0.9	12.0	12.0	50	74.0
1.0	10.0	9.5	40	59.5
1.1	8.0	6.5	30	44.5
1.2	6.0	6.0	24	36.0
1.3	5.0	4.5	18	27.5
1.4	4.5	3.5	14	22.0
1.5	3.5	3.0	11	17.5
1.6	3.0	2.0	8.5	13.5
1.7	2.3	1.5	6.5	10.3
1.8	2.0	1.5	5	8.5
1.9	1.6	1.0	4	6.6
2.0	1.4	1.0	3	5.4

* White and Schneidewind.(579)

than the same iron as cast but not pre-quenched. The higher the quenching temperature and the more rapid the rate of cooling (*i.e.*, water versus oil), the more rapid is the graphitization in both stages.

Schwartz* found that rapid cooling through the critical range by quenching in oil or water hastens the malleablizing process; the faster the rate of quenching, the greater is the acceleration of graphitization.

136. Short Annealing Cycles and the Resulting Properties.—Much work has been done by various investigators in the attempt to speed up the rates of graphitization in both first and

* U.S. Patent 1,688,438, Oct. 23, 1928.

second stages, *i.e.*, to produce "short-cycle" malleable. A recent summary of these investigations has been given by Highriter.(532) As long as the metallographic structure of the resulting product is the same as that of normal malleable, the properties of the short-cycle product do not appear to differ materially from those of normal malleable save for a slight tendency toward lower elongation.

Lauenstein* reported that the first-stage graphitization may be accelerated by first holding at 900°C. (1650°F.) for 4 hr. to precipitate widely distributed graphite nuclei, then raising the temperature to 1010°C. (1850°F.) to secure more rapid precipitation at this higher temperature.† The graphite formed at the higher temperature is thought to separate out adjacent to the large number of original nuclei rather than adjacent to the small number which would be formed were only the higher temperature employed, thus avoiding the poor mechanical properties which result from the presence of the larger graphite masses.

A special cycle, in which the temperature in the second stage is not continually held below the critical but is successively raised above and lowered below the critical in order to spheroidize any remaining cementite rather than to transform the last traces into temper carbon, is under commercial development. This more rapidly annealed product is said to meet the specifications of 70,000 lb. per sq. in. tensile strength, 50,000 lb. per sq. in. yield strength, with 15 per cent elongation, and regularly to exceed these figures; but there are few publications—except the report of Merz and Schuster, mentioned on page 373—which give technical details. It is understood that 0.5 to 1 per cent copper is used, which would tend, as Lorig and Smith(735) showed, to increase the yield strength and shorten the malleablizing time.

In addition to other modifications of the short-cycle process, there are quick-annealing processes in which some pearlite is retained; the products have much higher tensile and yield strengths than normal malleable, with correspondingly lower elongation. Little information is on record in scientific or technical publications by which the properties can be correlated with the structures and by which an appraisal might be made

* U.S. Patent 1,987,481, Jan. 8, 1935.

† A procedure very similar to this was suggested as early as 1925 by McNamara and Lorig.(165)

of the possibilities of specially annealed iron-carbon-silicon malleable iron as compared with alloyed malleable. It would be especially important to know the relation between ductility of these special products and their impact resistance.

A good picture of the possibilities and limitations of rapid processes may, however, be gained from the discussion of fundamentals by Sauveur and Anthony,[814] who made an interesting study of commercial malleable specimens, containing 2.20 to 2.50 per cent carbon, 0.85 to 1.00 per cent silicon, 0.26 to 0.37 per cent manganese, and 0.03 to 0.10 per cent sulphur, under various malleablizing schedules. The time and temperature used were shown in schematic graphs together with the resulting microstructure and properties.

The normal schedule, 4 to 36 hr. above the critical range (first stage), the time depending on temperature, followed by 24 hr. below the critical to spheroidize the pearlite and transform it to graphite, produced the normal completely graphitized structure with properties of 55,000 lb. per sq. in. tensile strength and 18.5 per cent elongation. Quenching after the first stage and reheating to 650°C. (1200°F.) for 12 hr. produced a similar matrix, but some of the graphite was in small nodules, which apparently had formed during reheating and had not collected upon the larger graphite aggregates which formed in the first stage. The properties were 59,000 lb. per sq. in. tensile strength and 19 per cent elongation.

Complete graphitization below the critical range, without any first-stage heating, which took from 10 to 15 days, resulted in an almost normal structure except for more perfectly rounded graphite nodules. Specimens showed 59,000 lb. per sq. in. tensile strength with 25 per cent elongation. Specimens carried through the first stage only and cooled at such a rate that the austenite transformed to pearlite but did not spheroidize had a tensile strength of 83,500 lb. per sq. in. with but 2.25 per cent elongation and with, of course, a pearlitic matrix.

A faster cooling from the first-stage temperature resulted in 116,000 lb. per sq. in. tensile strength and 3 per cent elongation. Quenching from the first-stage temperature and reheating to 650°C. (1200°F.) for 3 hr. produced similar structure and properties, but quenching after the first stage and reheating to 650°C. (1200°F.) for 5 hr. gave a spheroidized matrix with the interesting

combination of properties of 123,000 lb. per sq. in. tensile strength and 7 per cent elongation. The whole process can be accomplished in 40 hr. Similar structures can be obtained by cooling to 650°C. (1200°F.) without quenching and holding at that temperature long enough for spheroidization to take place. This paper is of importance as it emphasizes that the "quick-anneal" procedures are really a heat treatment of a high-carbon steel matrix in which are embedded free graphite particles resulting from the first-stage anneal above the critical range.

Merz and Schuster[639] studied the possibility of securing accelerated graphitization by successively raising and lowering the temperature to above and below the critical. The iron used contained 2.46 per cent carbon, 0.96 per cent silicon, 0.52 per cent manganese, 0.08 per cent phosphorus, and 0.047 per cent sulphur. Specimens 12 mm. (0.47 in.) round with 60 mm. (2.36 in.) gage were heated in an electric furnace to 1050°C. (1920°F.) for 3 hr. The temperature was then lowered to 700°C. (1290°F.) and immediately raised to 780°C. (1435°F.) and lowered again to 700°C. This was repeated at the rate of three cycles per hour for about 12 hr., after which the temperature spread was reduced to the range 710 to 760°C. (1310 to 1400°F.) and the heating and cooling were continued for 10 hr. longer. Specimens treated by this process contained 2.02 per cent total carbon, 1.97 per cent temper carbon, and 0.05 per cent combined carbon. Microscopic examination indicated that there was about 7.5 per cent of pearlite present. The average tensile strength was 62,500 lb. per sq. in., and the elongation in 60 mm. (2.36 in.) was 10.5 per cent. According to Merz and Schuster, speeding up the heating and cooling to four cycles an hour will result in complete graphitization.

It is obvious that it is not practical to attempt to secure rapid temperature changes in a commercial graphitizing furnace which, ordinarily, has a large thermal capacity and will, therefore, not heat up and cool down with great rapidity. Most short-cycle methods are, consequently, applicable to small-scale rather than large-scale practice. This fact should focus attention on quenching and tempering methods, as reported by Sauveur and Anthony,[814] as a method of accelerating the dissociation of cementite in white iron. Another factor which should be considered is the low elongation of much quick-annealed malleable.

Merz and Schuster[639] tried various methods of accelerating graphitization in addition to their repeated heating and cooling practice described above. Using various temperatures for the second stage of the cycle and times of 3 to 30 hr., they found that tensile strengths ranged from 57,000 to 84,000 lb. per sq. in., but elongation values were low, 1.3 to 6.1 per cent in 60 mm. (2.36 in.). The undissociated combined carbon ranged from 0.30 to 0.63 per cent, equivalent to 45 to 95 per cent pearlite in the structure.

137. Embrittlement of Galvanized Malleable Iron.—Ordinary malleable iron has appreciable, but not high, shock resistance. According to the Symposium on Malleable Iron Castings,[429] malleable iron at room temperature has a Charpy value of 7.5 or 8 ft-lb. and an Izod value of about 9 ft-lb. The corresponding values for ordinary gray cast iron are only 1 to 2 ft-lb.

As is well known, malleable castings which have been galvanized or heated to the galvanizing temperature of about 450°C. (840°F.) may become as brittle as cast iron, breaking with an intergranular fracture. The cause of this brittleness is not clearly understood, although it may be prevented by suitable heat treatment or by control of the composition. The brittleness is more marked in low-carbon, high-silicon alloys and increases with increasing phosphorus. It occurs frequently in the general run of normal high-strength malleable but is not so marked in cupola malleable. According to Bean,[97] an iron with 0.50 per cent silicon can have 0.26 per cent phosphorus, and one of 1.30 per cent silicon, 0.13 per cent phosphorus without being susceptible to galvanizing embrittlement, while Wolf and Meisse[429] put the limit at under 0.80 per cent silicon and 0.15 per cent phosphorus. Kelly[429] stated that embrittlement may occur even below these limits. Lorig and Smith[735] have shown that in the presence of sufficient copper the silicon and phosphorus limits can be extended without embrittlement.

The embrittlement phenomenon is not due solely to composition; for, with a given composition, the more slowly the material is cooled from the galvanizing temperature the less it is embrittled. As would be expected, quenched material is the most subject to embrittlement.

Marshall,[91,188] working at the National Bureau of Standards, found that, when annealed malleable iron is reheated to about

650°C. (1200°F.) and cooled preferably rapidly, subsequent heating to the galvanizing temperature no longer causes embrittlement. Moreover, a casting previously embrittled may be reclaimed by the subsequent higher temperature treatment. He also stated that the treated malleable retains its toughness at sub-zero temperature; the toughness does not fall off so rapidly with decreased temperature as in the case of untreated malleable, even if the latter is of a composition not subject to intercrystalline embrittlement. It would appear that some precipitation-hardening effect is involved as Kikuta[537] has pointed out, but just what the mechanism is has not been made clear. The phenomenon is of sufficient industrial importance so that with any proposal to produce an alloyed malleable it is necessary that the effect of the alloy upon galvanizing embrittlement be considered.

D. AUTHOR'S SUMMARY

1. Because malleable iron is stronger and much more ductile than ordinary gray iron, because it machines easily, and because it is more readily and economically cast into small intricate shapes than carbon steel, malleable castings are widely used for agricultural implements, automotive and railroad equipment, pipe fittings, hardware, machine tools, electric and industrial power equipment, and for many other applications.

2. There are two processes for the manufacture of malleable iron. In the European white-heart process castings of white iron are annealed in a packing of iron oxide, which results in complete decarburization. The structure of white-heart malleable is almost wholly ferritic. In the black-heart process, used almost exclusively in the United States and to an increasing extent abroad, white-iron castings are annealed in an almost inert or a weakly oxidizing packing. The castings may be decarburized at the surface. Most of the cementite, however, is decomposed, and the carbon remains in the casting as small, round nodules of graphite (temper carbon). The structure of black-heart malleable consists of ferrite containing numerous islands of temper carbon.

3. White-heart castings have about the same tensile and yield strengths as black-heart castings, but the elongation

(2 to 6 per cent) of the former is much lower than that of the latter (10 to 30 per cent).

4. White iron to be malleablized by the black-heart process is melted in an air furnace, cupola, electric furnace, or occasionally in a small open hearth. Most of the malleable iron produced in the United States is melted in the air furnace, but a considerable tonnage, for pipe fittings, is melted in the cupola. The other two methods are used to a very limited extent.

5. The composition of malleable is adjusted according to the size of the casting and according to the properties desired. The carbon content may vary from 2.0 to 3.0 per cent, and the silicon, which varies inversely with the carbon, from 1.0 to 0.7 per cent. As much as 0.2 per cent phosphorus, 0.25 per cent sulphur, and from 0.3 to 0.6 per cent manganese may be present. Low carbon percentages are used where strength and ductility are major considerations, 2.8 to 3.0 per cent when prevention of foundry troubles is a major item, and 2.3 to 2.6 per cent for a fair combination of all desired characteristics.

6. The best available data for such physical properties as thermal expansion, mean specific heat, and electric resistivity are summarized. These may be useful as first approximations to the true values. Data are also quoted which indicate that, although machinability of malleable iron cannot be dealt with quantitatively since there is no standardized test nor unit of measurement, as measured by tool life or cutting speed in commercial operations, malleable iron machines more readily than any other ferrous material. Welding is not commercially feasible.

7. Typical tensile properties of cupola malleable are: tensile strength, 48,000 lb. per sq. in., yield strength, 34,000 lb. per sq. in., and elongation in 2 in., 9 per cent. For an average of 20,000 heats of air-furnace malleable the corresponding values are 54,000 lb. per sq. in., 36,000 lb. per sq. in., and 18 per cent. Despite the fact that cupola malleable has slightly lower tensile strength and much lower elongation, fittings made of this material have a higher bursting strength, which is especially pronounced in the smaller sizes, than fittings made from air-furnace malleable.

8. Malleable iron is unique among ferrous materials in that the percentage elongation increases with tensile strength. If tensile-strength variations are caused by original composition,

an increase of 5000 lb. per sq. in. tensile strength is accompanied by an increase of about 8 per cent elongation. If these variations are caused by factors other than composition, the same increase in tensile strength is accompanied by an increase of about 3 per cent elongation.

9. The tensile properties of malleable iron are affected, though in some cases not importantly, by surface decarburization, quick cooling in freezing, roughness of the surface, size of the section, method of gating, and completeness of the anneal.

10. In the past 15 years, the average tensile strength of black-heart malleable iron has increased from about 49,000 lb. per sq. in. to 55,000 to 59,000 lb. per sq. in., and the elongation from 10 to around 20 per cent. Most of this improvement has been caused by lowering the average carbon content from about 2.5 to around 2.1 per cent and by an accompanying careful control of melting and casting practice.

11. The graphitizing cycle for black-heart malleable consists of two stages. In the first, the white-iron castings, packed in pots, are heated slowly (about 2 days for the usual charge) into the austenite range, 845 to 925°C. (1550 to 1700°F.), and held there for 48 to 60 hr. until the excess cementite is graphitized. The second stage consists in cooling the charge very slowly from the first-stage temperature. If cooled slowly enough—say at 5°C. (10°F.) per hr.—graphitization of the cementite keeps pace with the precipitation of the carbon from the austenite as the material cools from the first-stage temperature to the critical range. In cooling through this range, gamma iron transforms to alpha iron, and the remaining cementite is graphitized. Sometimes the second stage is accomplished by cooling the charge fairly rapidly from the first-stage temperature to just below the critical range and holding at this temperature until graphitization is complete.

12. The heating and cooling rates and the time of holding at the graphitizing temperature depend for any particular charge on the composition, especially upon the silicon percentage. In general, the total time for graphitizing (for the whole cycle) decreases with increasing silicon.

13. Much work has been done in an effort to shorten the time for the total graphitizing cycle and to produce a "short-cycle" malleable iron with satisfactory properties. Among the

suggested methods for shortening the cycle are: (*a*) holding at 900°C. (1650°F.) and then raising the temperature to 1010°C. (1850°F.) to secure more rapid precipitation of carbon; (*b*) using a special treatment for the second stage which consists of successively heating above and cooling below the critical range to spheroidize rather than graphitize the last traces of cementite; and (*c*) methods involving quenching and tempering. One of the outstanding difficulties in shortening the time for the graphitizing cycle is securing rapid temperature changes in the large commercial graphitizing furnaces.

14. Malleable castings which have been galvanized or heated to the galvanizing temperature of about 450°C. (840°F.) frequently become as brittle as cast iron. The cause for this is not known, but it may be prevented by a suitable heat treatment or by adjusting the composition. The brittleness occurs more frequently in air-furnace iron than in the product of the cupola and is more pronounced in low-carbon high-silicon alloys than in the high-carbon low-silicon grades. It increases with increasing phosphorus. It is not due solely to composition; the more slowly material of a given composition is cooled from the galvanizing temperature the less it is embrittled.

CHAPTER XI

BEHAVIOR OF IRON-CARBON ALLOYS UNDER REPEATED STRESS

Methods of Testing—Endurance Limit of Wrought Iron-carbon Alloys—Endurance Limit of Cast Iron-carbon Alloys—Factors Affecting Endurance Limit—Corrosion Fatigue and Protective Coatings—Author's Summary

Since many failures of metals subjected to repeated stress occur under loads not only far below their tensile strength but even below their yield strength, the problem of finding the stress below which the metal will not fail under long-continued repetition of loading has become highly important. Wöhler, in Germany, made pioneer investigations in the field between 1850 and 1870, as did Fairbairn, in England, in the 1860's.

Most of the modern information on "fatigue" or "endurance" of metals, as the properties under repeated stress are commonly termed, has been made available during the past 15 years, notably through the publications of Moore[(224)] and coworkers at the University of Illinois, of McAdam[(472)] at the Naval Experiment Station and the National Bureau of Standards, of Gough[(178)] of the National Physical Laboratory, England, and of the Research Committee on Fatigue of Metals of the American Society for Testing Materials,[(284)] as well as many others.

A. METHODS OF TESTING

It is only within the past decade that the necessary precautions with respect to shape of endurance test bars, radius of fillets between breaking section and grips, and the proper polishing of the specimen before the test, necessary to secure reliable and reproducible results, have been fully understood and adopted. With specimens that conform to these requirements a wide variety of testing machines may be used with comparable results.

138. Preparation of Specimens.—A fatigue specimen must be so designed and so loaded that the location and magnitude of the

maximum stress can be accurately determined. This necessitates the use of specimens having generous fillets. When a straight reduced portion meets a fillet, great care must be exercised to produce a smooth junction, else there will be distinct local concentration of stress. Smoothly radiused bars without a straight portion are more generally used at present.

When it is desired to obtain the actual fatigue properties of a material, samples are carefully polished; the recommended procedure is to polish the samples in such a manner that the scratches are parallel to the direction of the stresses. Of course, data on polished samples do not necessarily give the desired information regarding the behavior of the material in service, and it is entirely in order to test samples having a surface similar to that found in actual parts. The condition of the surface or the nature of the layer of metal adjacent to the surface has a very pronounced effect on the endurance limit of specimens stressed by bending but has less effect on specimens subjected to axial loading. Hence for proper appraisal of recorded data the condition of the surface must be stated.

139. Types of Loading.—Two methods of loading are chiefly used, the more common being the rotating-beam method in which portions of the specimen are alternately subjected, as the specimen rotates, to fiber stresses varying from a given stress in tension to the same stress in compression; with the other method the specimen is subjected to direct push and pull, *i.e.*, to axial loading (usually in the Haigh machine), and the load may be varied about a zero mean value between the same tensile and compressive loads, thus using the same stressing cycle as the rotating-beam test, or it may be varied about some other mean.

While interesting engineering data are obtained by other than completely reversed stressing, from the point of view of evaluation of materials the results of balanced loading about a zero mean will be suitable for the purposes of this monograph. Unless otherwise noted, all data cited are from rotating-beam tests and refer to completely reversed stresses. Values of stress refer to the "half-stress range" and are the values of maximum tensile or compressive stress produced in the sample.

The usual rotating-beam specimen, with its smoothly radiused change of section to a minimum at the middle, applies the

maximum load practically only to a line on the circumference of the bar, since the stress falls off below the surface and is zero at the neutral axis. The tapered cantilever bar, as used by McAdam,[472] stresses more of the surface uniformly. Because of the rapid fall in stress below the surface of either type of rotating-beam specimen, only a very minute volume of metal is, in any case, tested at the maximum stress, and chance may or may not locate "stress raisers" (such as surface scratches or interior discontinuities which increase the local stress at such points over that calculated to be present when such discontinuities are disregarded) in that volume. Axially loaded specimens, however, have a much larger equally stressed volume, and one such test bar is equivalent in volume tested to a large number of rotating-beam bars. The axially loaded specimen fails when a crack starts at any point. Where 100 rotating-beam bars tested at the same calculated stress would fail over a range of stress reversals and would probably yield a normal-distribution curve, axially loaded bars will tend to break at the low-cycle end of the distribution curve given by the rotating-beam specimens. Hence, it is to be expected that axial-loading tests, properly conducted with respect to true axiality of loading and the other necessary precautions, would give lower endurance limits than rotating-beam tests made on but a few specimens whenever the material tested is non-uniform and contains local stress raisers. When the material is strictly homogeneous, so that the rotating-beam tests do not show scatter, the axial test should give an endurance limit as high as the rotating beam.*

With polished rotating-beam specimens most observers do not find an appreciable size effect when small- and large-diameter specimens are compared. Faulhaber, Buchholtz, and Schulz,[609] however, while admitting that specimens with an ideal polish should show no size effect, found such an effect in poorly finished bars and, in common with other observers, found a notable effect when definite notches were present, with a more marked effect in hard than in soft steels. They pointed out that the volume at high stress is greater in large-diameter bars than in small ones. In the larger rotating-beam bars, there is, for this reason, a greater chance for the presence of stress raisers within the

* McAdam (private communication) doubts this.

volume so highly stressed that the actual stress about the stress raiser is above the endurance limit, as in the case of the axially loaded bar.

A statistical effect, which tends to show proportionally lower endurance limits and endurance ratios (ratio of endurance limit to tensile strength) when greater volumes of material are tested, was shown in a series of tests by Peterson(400) on a steel containing 0.44 per cent carbon and having a tensile strength of 81,500 lb. per sq. in. Specimens of this steel 1 or 2 in. in diameter gave an endurance ratio of 0.39, specimens 0.5 in. in diameter a ratio of 0.41, and specimens with a diameter of 0.25 or 0.05 in. a ratio between 0.43 and 0.44. A steel containing 0.42 per cent carbon and having a tensile strength of 74,000 lb. per sq. in. had an endurance ratio between 0.43 and 0.46 for specimens varying in diameter from 0.05 to 1 in., and a 0.57 per cent carbon steel with a tensile strength of 102,000 lb. per sq. in. gave an endurance ratio between 0.47 and 0.48 for specimens from 0.05 to 0.5 in. in diameter.

140. Relative Results Obtained by Rotating-beam and Axial-loading Methods.—There has been much discussion in the literature regarding the relative results of the rotating-beam and axial-loading tests, but recent careful work by France(450) seems to settle the question. On a wide range of materials—open-hearth iron with 0.02 per cent carbon and steels containing from 0.30 to 0.90 per cent carbon—he found no case where the axially loaded specimens gave a higher endurance limit than the rotating-beam specimens. Open-hearth iron and quenched and tempered steels containing 0.45 and 0.87 per cent carbon gave practically identical values by the two methods, but the endurance limit of the annealed steels as determined by axial loading was only 75 per cent of the limit obtained by the rotating-beam test. With another steel, containing 0.47 per cent carbon, in the annealed condition the endurance limit for axial loading was 80 per cent of the limit obtained by the rotating-beam test; in the quenched and tempered condition the ratio was 85 per cent instead of 100 per cent obtained with the two other steels in the quenched and tempered condition.

A recent discussion by Gough and Sopwith(615) dealt with reasons why the endurance limit determined by axial loading is frequently lower than the limit determined by reversed flexure.

One might conclude that the same factors which tend to produce a low endurance ratio with the rotating-beam tests on annealed steels also tend to make annealed steels show lower endurance limits for axial loading than with the rotating-beam test. McAdam* thinks it possible that even in uniform metals, those with relatively low yield strength and high ductility, endurance limits by rotating-beam tests may be higher than by axial-loading tests (see also McAdam and Clyne[739]).

141. Accelerated Endurance Tests.—Since endurance tests must be carried to so many million cycles, there is a great urge to run the tests at a high stress-cycle frequency so as to make them require less time. It is, therefore, necessary to inquire whether the endurance limit will be higher or lower at different frequencies and whether, if there is any speed effect, it would be different in different steels, so that one may know whether a test, carried out at a high frequency, gives reliable data for application to service at lower frequency. The rotating-beam machines which use large specimens and hence heavy loads develop bearing troubles at very high speeds. Krouse[730] has designed a successful high-speed rotating-beam machine using a rather small specimen.† Gill and Goodacre[701] have described a rotating-beam apparatus for high-speed testing of small wire. Other investigators[309,689] have utilized reversed-bending machines in which the specimen is driven magnetically or made to vibrate by air jets. Some of these have reported speed effects, but, as has been pointed out,[370] when notable speed effects have been found, the test conditions were such as to make the matter of

* Private communication.

† A new high-speed machine has been recently developed by R. R. Moore, Senior Metallurgist, U. S. Naval Aircraft Factory, Philadelphia, Pa. This is a rotating-beam type similar in operation to the previous R. R. Moore machine, but operating at a speed of 10,000 r.p.m. instead of 1750 as did the previous machines and the old Farmer machines. The new machine uses the same specimen as the previous machines so that specimens are entirely interchangeable. This interchangeability, of course, does not hold for the Farmer machines which have been almost entirely abandoned since the development in about 1925 of the R. R. Moore machine using the short specimen. The increase in speed makes one of the new machines equal to six of the previous machines. A test on steel to the required 10,000,000 cycles for locating the endurance limit, which formerly required about 100 hr. to run, can be completed on the new machine in about 16 hr., which is less than one working day for a fatigue machine.

accurate stress measurement a very difficult one. In some cases internal friction may heat the test specimen and thus affect the results somewhat, though, as is pointed out in the next chapter, most steels do not have their endurance limits greatly altered by a moderate increase above room temperature. No reliable conclusions on speed effect can be drawn from those test methods whose true stress values are doubtful. Krouse's data indicate that there may be very slight speed effects, but, if they exist, they seem to be of very low order of magnitude. Results obtained by any method of testing whose manner of stress measurement is beyond reproach will probably not be vitiated by the speed effect.

142. Torsional-fatigue Tests.—Torsional-fatigue tests are occasionally made, but relatively few data on the results of such tests are available. In general, their indications are much like those of the rotating-beam tests, but the actual magnitude of the torsional endurance limit is smaller. As a rough approximation it may be assumed that the endurance limit in reversed torsion (shear) is between 50 and 60 per cent of the tension-compression endurance limit determined by the rotating-beam method. Hankins,[257] Moore,[224] Gough,[178] and McAdam[60,131] gave values of torsional endurance limit ranging from 48 to 64 per cent of the rotating-beam limit. Mason[45] carried out extensive work on torsional fatigue. Far too few data are available to warrant generalizations regarding the effects of carbon or structure on this ratio.

Moore and Picco[742] pointed out that under torsional stress a brittle material like cast iron does not fail in shear but under diagonal tension, so that endurance limits in flexure and torsion are about the same on such materials. They showed that the endurance limit in completely reversed torsion runs from 78 to 95 per cent of that in completely reversed bending (flexure of a flat specimen) for a group of cast irons whose tensile strength ranged from 44,000 to 76,000 lb. per sq. in. The endurance limit in flexure ran from 19,000 to 25,000 lb. per sq. in. and that in torsion from 16,000 to 22,000 lb. per sq. in. In this connection, McAdam pointed out*

that many metals other than cast iron fail under diagonal tension. This does not necessarily imply that the torsional fatigue limits in

* Private communication.

flexure and in torsion should be the same. In tension there is compressive stress at an angle of 45 deg. to the tensile stress, consequently the specimen is under combined mutually perpendicular tensile and compressive stresses. This is not the case in a rotating-beam test.

143. Plotting and Interpreting Data.—The endurance limit is the half-stress range, given in the United States as pounds per square inch, which a specimen initially subjected to that repeated stress will continue to endure for an indefinitely long period. It is determined by testing a series of like bars, each under a different load, and plotting the number of cycles which cause fracture against the half-stress range, using a smaller load for each new specimen until a load is reached at which the specimen does not break in 5 or 10 million cycles. Except in the case of very hard steels, it may be said that, if a steel specimen endures 10 million cycles of a certain repeated stress, it will continue for an indefinite period to withstand further application of that same stress or a lower stress. With quenched steels and those tempered at low tempering temperatures, fractures may be met at 50 or 100 million cycles, but even in such steels the load needs to be reduced but little below that for a life of 10 million cycles before indefinite life is assured.

While there is some doubt regarding the existence of a true endurance limit in the case of certain non-ferrous metals and alloys (which may break after hundreds of millions or even billions of cycles, so that the endurance properties for them have to be reported in terms of the number of cycles used), with steels there appears to be a true endurance limit, expressible in lb. per sq. in. Equally definite endurance limits may exist for all non-ferrous alloys, but in the cases where it requires several runs of many billions of cycles to establish whether or not the curve becomes truly horizontal, evidence is naturally slow in accumulating.

The endurance limit is obtained by plotting the stress, S, against the number, N, of cycles producing failure, the endurance limit being that stress at which the S-N curve is parallel to the N axis. On account of the large number of cycles involved in testing, the number of cycles, N, is plotted on a logarithmic scale. Stress may be plotted on a logarithmic scale or not, but it is ordinarily not so plotted. Some fatigue data reported by Moore and Kommers[76] are plotted by the semilogarithmic method in Fig. 125. The data are for an ingot iron in the hot-

rolled condition and for normalized steels containing 0.37, 0.52, and 1.20 per cent carbon.

The exact shape of the high-stress end of the S-N curve varies somewhat with the type of specimen and the manner of loading; the curves may or may not show a sharp "knee" at the endurance limit. While very interesting facts may be learned from the shape of the S-N curve when enough specimens are tested to establish its course exactly (and more attention is being paid to this), in many reports only endurance limits are given and no

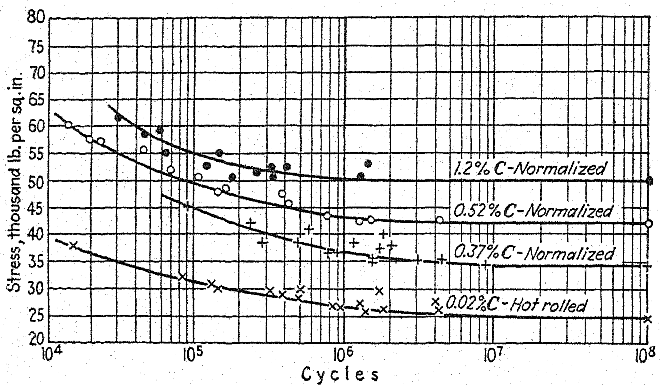

FIG. 125.—Typical stress-cycle curves. (*Moore and Kommers.*[76])

mention is made of the shape of the curve itself. Subtle distinctions among steels that may be detected from the shape of the curve cannot yet be broadly discussed owing to limited information. The following discussion is, therefore, primarily confined to the facts brought out by the mere value of the endurance limit.

Some lots of steel show wide scatter of points on the S-N curve and, when a sufficiently large number of specimens is tested, most steels show a band rather than a line. Were it not for the tediousness of endurance testing, it would perhaps be the most illuminating method of evaluating uniformity, homogeneity, and "quality" of steel that is available. The width of the scatter band is a good indicator of the presence of stress raisers.

B. ENDURANCE LIMIT OF WROUGHT IRON-CARBON ALLOYS

In the discussion of endurance limits given hereafter emphasis is placed on the ratio of endurance limit to tensile strength rather

than on actual endurance values. In addition to the relation of endurance limit to other properties, the variables which affect the endurance limit are also discussed. These include heat treatment, cold work, composition, and the effect of the direction of prior hot working.

144. Relation between Endurance Limit and Other Properties. An extremely convenient approximation (but only an approximation) to a constant ratio between endurance limit and tensile strength holds for sound ferrous alloys; the endurance limit is roughly half of the tensile strength. This holds for carbon and alloy steels of all carbon contents, both cast and wrought steels, for pearlitic, sorbitic, and troostitic steels, as well as for austenitic steels, cast iron, and malleable iron. It also holds for cold-worked steels. Only in the case of very hard steels, such as those made essentially martensitic by quenching and which are not tempered or which are given only a low-temperature tempering treatment, is there a notable deviation from this rule.* From a soft cast iron with a tensile strength of 20,000 lb. per sq. in. or an ingot iron with a tensile strength of 45,000 lb. per sq. in. up to a hardened alloy steel with a tensile strength of 250,000 lb. per sq. in. the endurance limits average 50 per cent of the tensile strength or very slightly less and the extremes of variation very rarely fall outside of the limits 35 and 65 per cent. There are, however, no well-established endurance limits of over 125,000 lb. per sq. in. on record, and there is a tendency in cast materials for the ratio to run a bit lower than 50 per cent, perhaps more closely approximating 40 per cent. The approximate ratio of 50 per cent also holds for tests made at very low and at quite high temperatures. The engineer should probably use 40 per cent of the tensile strength, where he cannot obtain actual endurance-limit data, as a safer approximation of the probable endurance limit to cover a greater proportion of the deviations on the low side, and to this approximate figure he should, of course, apply a factor of safety. Where design needs to be made on the endurance-limit basis, actual tests should be carried out.

The proportionality between endurance limit and tensile strength does *not* hold well enough to remove the necessity for

* McAdam (private communication) believes that there may be a deviation (a lower ratio) from this rule for annealed carbon steels or even for quenched and tempered low-carbon steels.

endurance testing, for a steel having a tensile strength of 150,000 lb. per sq. in. may have an endurance limit as low as 55,000 lb. per sq. in. or as high as 95,000 lb. per sq. in. A definite relationship between endurance limit and tensile strength is not found with non-ferrous materials, except possibly for nickel-chromium alloys, and no good theoretical explanation has been adduced for it in ferrous alloys.

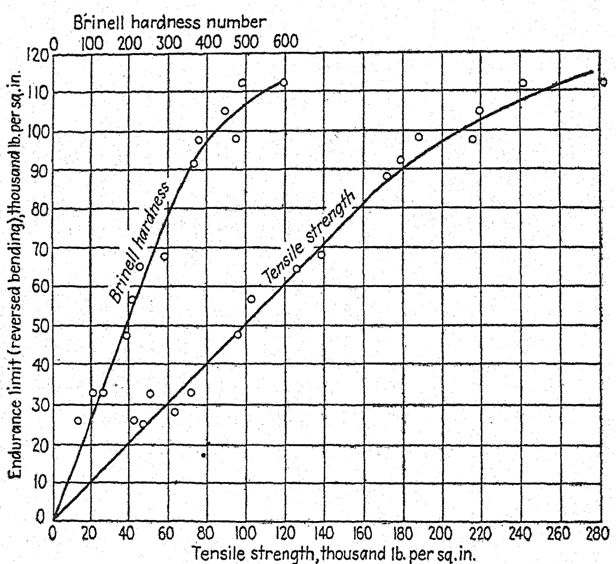

FIG. 126.—Approximate relation between endurance limit, tensile strength, and Brinell hardness. (*Moore.*[326])

Since hardness has an approximately straight-line relationship to tensile strength, the correlation of endurance limit and hardness is equally good. Multiplying the Brinell hardness by 250 gives a rough approximation to the probable endurance limit of steel. Figure 126, after Moore,[326] shows these relationships. Houdremont and Mailänder[307] gave a distribution curve based on results from 170 steels of different compositions and treatments, showing how uniformly the results group themselves about the 50 per cent endurance ratio.

There is no direct correlation between endurance limit and other properties of steel, such as ductility, though ductility quite definitely plays a part in the life of a steel section in severe

service, since the more ductile steels often have better resistance to overstress or shock loading, which they may also have to endure in actual service. From mathematical and other considerations it can be shown that the local stress at a hole in a plate, for example, may be 3 times the average stress; yet endurance tests of specimens containing holes or sharp notches seldom, if the material is ductile, show a threefold weakening. In this connection, Scorah[657] compared the endurance limits of standard endurance specimens of heat-treated steels with those of specimens of the same minimum diameter but having a shoulder of larger diameter filleted down to the shaft by different radii. The stress-concentration factors varied from about 1.2 in a specimen with a shoulder of small diameter joined by a fillet of large radius in a relatively soft steel up to the theoretical 3 in a specimen with a shoulder twice the diameter of the shaft joined by practically no radius at all, *i.e.*, a sharp shoulder, in a hard steel. Only by such extremely abrupt changes in section, or the sharpest of notches, does the endurance limit fall as much as mathematical theory would indicate. That the actual performance is usually better than the theory would indicate is probably due to the ductility and possibly to the work-hardening capacity of the material.

Various attempts have been made to find a better correlation between endurance limit and some complex function of several properties (always including tensile strength) than now exists between endurance limit and tensile strength alone. These proposed complex formulas frequently attempt to correct for the deviation of the endurance ratio from a constant value by introducing additive or subtractive factors depending upon yield strength, elongation, and reduction of area. Some of these formulas appear to be better within limited ranges of values or on specific types of steel, but they are not satisfactory for all ranges or for all types.

The simple endurance ratio, the endurance limit divided by the tensile strength, is probably the most useful index of endurance qualities of ferrous materials.*

* McAdam (private communication) would not limit this usefulness to ferrous materials. "Considering classes of non-ferrous metals, it can be applied just as well. For example, a normal endurance ratio of 25 per cent can be used for heat-treated duralumin."

145. Variation of Endurance Limit with Heat Treatment.—Steels quenched and not tempered, or tempered at very low temperatures and thus left in a very hard condition and presumably containing high internal stresses, may show low endurance ratios. Moore found a ratio of 38 per cent for a quenched and untempered steel containing 0.41 per cent carbon and 3.5 per cent nickel; the Brinell hardness was 520 and the tensile strength 294,000 lb. per sq. in. Lessells[216] studied brine-quenched specimens of steels containing 0.42 and 1 per cent carbon. The quenched steel, containing 0.42 per cent carbon, after being heated in boiling water for 1 hr. had a Brinell hardness of 578, a tensile strength of 230,000 lb. per sq. in., and an endurance limit of 81,000 lb. per sq. in. The endurance ratio was, therefore, only 35 per cent. From length changes resulting when surface layers were removed by grinding, it was estimated that the material tempered merely by heating in boiling water contained a residual stress of 110,000 lb. per sq. in. On tempering at 310°C. (590°F.) for 1 hr. the Brinell hardness fell to 477 and the tensile strength to 215,000 lb. per sq. in., but the endurance limit rose to 98,000 lb. per sq. in., yielding an endurance ratio of 48 per cent. The residual stress fell to 30,000 lb. per sq. in. The 1 per cent carbon steel tempered at 300°C. (570°F.) had a Brinell hardness of 495, a tensile strength of 241,000 lb. per sq. in., an endurance limit of 112,000 lb. per sq. in., an endurance ratio of 47 per cent, and no residual stress. At very high hardness and strength, the endurance ratio may thus be lower than is indicated in the average curves of Fig. 126 if internal stress is present. The increase in endurance limit of a water-quenched 0.87 per cent carbon steel on tempering at 290 and 425°C. (550 and 800°F.) is shown by French[612] in Fig. 127.

It is known that compressing the surface of an endurance specimen by actual cold working raises the endurance limit. Bühler and Buchholtz[596] pointed out that quenching from 600°C. (1110°F.) (below A_1) introduces a compressive stress at the surface of a specimen, with a corresponding tensile stress at the core; they studied three steels, one of about 0.30 per cent carbon, one of 0.57 per cent carbon, and a medium-manganese steel, to find if such stress would raise the endurance limit. By quenching in ice water from 600°C. (1110°F.) they introduced from 35,000 to 48,000 lb. per sq. in. compressive stress at the

surface and found the endurance limit in reversed bending to increase by 3000 to 8500 lb. per sq. in.

After 8,360,000 cycles at the endurance limit the initial compressive stress at the surface was found to be decreased practically to zero, while some compressive stress still existed below the surface, but less than originally existed at that point. The tensile stress at the core was also diminished.

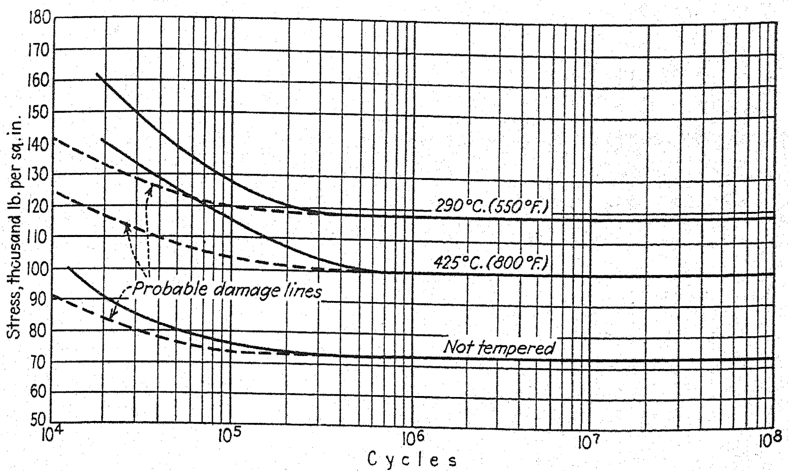

FIG. 127.—Effect of tempering temperature on the endurance limit and susceptibility to damage by overstress of a steel containing 0.87 per cent carbon, 0.28 per cent manganese, and 0.28 per cent silicon, water quenched from 760°C. (1400°F.). (*French*.[612])

146. Influence of Cold Work on Endurance Limit.—As mentioned above, hardening induced by cold working raises the endurance limit, and the increase is approximately proportional to the increase in tensile strength and hardness. Aitchison[64] reported that a cold-drawn steel containing 0.25 per cent carbon had a tensile strength of from 80,000 to 92,000 lb. per sq. in. and an endurance ratio of 47 per cent; after "blueing" by heating within the range 250 to 550°C. (480 to 1020°F.), a range within which the tensile strength is unaffected or is decreased by 10 per cent, the endurance ratios varied from 45 to 49 per cent.

The effect of cold working on the endurance ratio of a 0.18 per cent carbon steel is shown by data reported by Moore and Kommers.[76] As received, the steel had a tensile strength of 61,-500 lb. per sq. in. and an endurance ratio of 45 per cent. After

stretching so that the diameter was reduced from 0.50 to 0.48 in. (tensile strength 67,500 lb. per sq. in.) the endurance ratio was 52 per cent, and after stretching until the diameter was 0.44 in. (tensile strength 73,500 lb. per sq. in.) the ratio was 56 per cent.

The endurance properties of cold-worked and annealed screw-machine stock were also studied by Moore and Kommers.[76] The steel contained 0.20 per cent carbon, 0.67 per cent manganese, 0.025 per cent phosphorus, 0.09 per cent sulphur, and 0.03 per cent silicon. In the cold-drawn condition it had a tensile strength of 87,000 lb. per sq. in. and an endurance ratio of 47 per cent. After a stress-relieving anneal at 705°C. (1300°F.) the endurance ratio was 47 per cent, and after a full anneal at 815°C. (1500°F.) the endurance ratio was 43 per cent.

Severe cold working, such as "overdrawing," which tends to produce actual or incipient internal fractures or high internal stress, not relieved by tempering would be expected to affect adversely the properties of steel as it has been observed in non-ferrous metals.[132,522] Templin,[664] however, failed to find evidence of overdrawing in severely worked aluminum alloys. Dowling and coworkers[689] found an endurance ratio of only about 0.25 on very hard-drawn steel wire; an anneal at 450°C. (840°F.) left the endurance limit practically unaffected, although it reduced the tensile strength more than 15 per cent.

In a comprehensive study of the fatigue properties of medium- and high-carbon wires drawn from patented (lead- or air-quenched) rods Gill and Goodacre[701,797] investigated the effect of carbon content, the effect of surface conditions, whether the limiting fatigue stress could be used to determine the point where the wire becomes overdrawn, the effect of reheating at low temperatures on the fatigue limit, and other factors. Some of these variables are discussed in the section on the effect of surface conditions (page 409). In regard to overdrawing these investigators found that "under certain conditions the fatigue properties give an indication of overdrawing, but no generalization is possible at present."

Gill and Goodacre found further that drafting patented medium- and high-carbon rods increases the tensile strength and endurance limit by about the same relative amount; the ratio thus remains practically the same. This is shown by the following endurance ratios:

Reduction by drafting, per cent	Endurance ratio of lead-patented wire		
	0.36 % carbon	0.55 % carbon	0.79 % carbon
25	0.34	0.31	0.27
50	0.37	0.31	0.28
75	0.35	0.30	0.26
80	0.34	0.30	0.26
85	0.33	0.27	0.26
90	0.31	0.27	0.26

The endurance tests were made according to the method described on page 411, on wire free from decarburization but with a surface as it came from the dies. Polishing would have raised these ratios by about 25 per cent as is shown by a series of tests on the 0.79 per cent carbon lead-patented wire drawn 72.5 per cent:

Condition of Surface	Endurance Ratio
Not polished	0.24
Machine polished	0.29
Polished with 000 emery	0.30

In the second half of the investigation,[797] it was found that in general there is an increase in the limiting fatigue stress when medium- or high-carbon wire drawn from patented rods which are free from decarburization is reheated to about 150 to 200°C. (300 to 390°F.). There is, however, a critical reduction by drafting at which point, under certain tempering conditions, the limiting fatigue stress may fall considerably below that for the as-drawn specimen. This critical reduction by drafting becomes progressively lower as the carbon content increases. Some of Gill and Goodacre's results are given in Table 80. The critical reduction in drafting which is accompanied by a lower endurance ratio for the tempered wire is about 80 to 85 per cent for the 0.55 per cent carbon steel and about 75 to 80 per cent for the 0.79 per cent carbon material. In the latter the ratio, for reductions of 75 and 80 per cent and for tempering temperatures of 150 to 300°C. (300 to 570°F.), is actually lower than for the corresponding untempered wire. It should be noted that all of the values given in Table 80 were obtained on wire as

TABLE 80.—EFFECT OF AMOUNT OF DRAFTING AND SUBSEQUENT TEMPERING ON ENDURANCE RATIO OF STEEL WIRE*

Reduction in drafting, per cent	0.55 per cent carbon					0.79 per cent carbon				
	Not tempered	Tempered 20 min. at				Not tempered	Tempered 20 min. at			
		150°C. (300°F.)	200°C. (390°F.)	300°C. (570°F.)	400°C. (750°F.)		150°C. (300°F.)	200°C. (390°F.)	300°C. (570°F.)	400°C. (750°F.)
25	0.31	0.38	0.37	0.35	0.41	0.27	0.26	0.26	0.28	0.29
50	0.31	0.38	0.33	0.38	0.41	0.28	0.33	0.33	0.32	0.36
75	0.30	0.37	0.38	0.36	0.41	0.26	0.24	0.23	0.25	0.26
80	0.30	0.32	0.32	0.34	0.39	0.26	0.23	0.23	0.23	0.27
85	0.28	0.30	0.34	0.31	0.37	0.26	0.32	0.31	0.29	0.31
89	0.27	0.27	0.32	0.35	0.39	0.26	0.30	0.29	0.29	0.32

* Gill and Goodacre.(797)

it came from the die—unpolished—and so likely to contain minute die marks and other surface imperfections.

Kommers[384] compared hot-rolled ingot iron with a tensile strength of 44,500 lb. per sq. in., an endurance limit of 26,000 lb. per sq. in., and an endurance ratio of 59 per cent with the same material as cold drawn, which had a tensile strength of 73,000 lb. per sq. in., an endurance limit of 33,500 lb. per sq. in., and an endurance ratio of 46 per cent. The tensile strength was increased 65 per cent and the endurance limit but 27 per cent. McAdam* suggested that the specimen may have contained considerable internal stress. Kommers considered that, since the endurance limit of the hot-rolled ingot iron was 3500 lb. per sq. in. above its yield strength, the material can be improved by cold working up to the point where incipient cracks and internal stresses are set up.

147. Influence of Composition on Endurance Limit.—In absolute values, endurance limits naturally increase with the tensile strength as carbon increases in normalized and annealed steels. However, the endurance ratio may not be constant. Crook[292] calculated from the data of Gough[178] and others that the endurance ratios for steels with different ranges of carbon, after annealing or normalizing and after quenching and tempering, are approximately as follows:

Carbon, Per Cent	Endurance Ratio, Per Cent
Annealed or Normalized	
0.10 to 0.35	45
0.36 to 0.65	43
0.66 to 1.20	35
Quenched and Tempered	
0.10 to 0.45	40 to 43
0.50 to 0.65	53
0.66 to 1.20	48

He stated that endurance limits calculated from these ratios may be in error from 2 to 8 per cent.

Houdremont and Mailänder's[307] data, summarized in Fig. 128, are in agreement with McAdam's[104] conclusion that:

> The endurance ratio of ingot iron is higher than that of any pearlitic carbon steel. The ratio decreases with increase in pearlite. Experi-

* Private communication.

ments with annealed steels of about 0.10 per cent carbon indicate that not only the endurance ratio but even the endurance limit is lowered by the first addition of cementite to ferrite. The raising of the endurance limit by increase of cementite percentage does not keep pace with the increase in tensile strength. . . . A quenching and tempering treatment applied to low- and medium-carbon steels decreases the size of the carbide particles and thus raises the endurance limit. By such treatment of the lower carbon steels the endurance ratio is not greatly

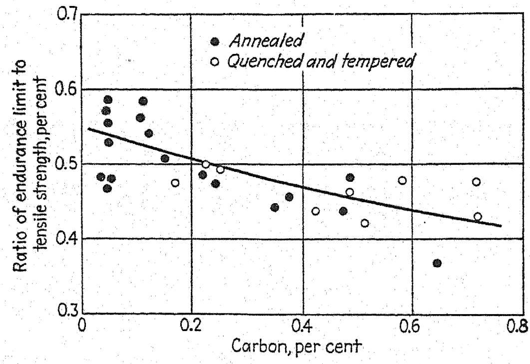

FIG. 128.—Effect of carbon content on endurance ratio. (*Houdremont and Mailänder.*[307])

increased. With increase in carbon content, however, quenching and tempering increases the endurance ratio as well as the endurance limit.

From the data reported by Gill and Goodacre,[701] obtained on medium- and high-carbon wires drawn from patented rod, and free from decarburization, it appears that an increase in carbon results in a decrease in the endurance ratio. This is shown by the following table of ratios:*

Carbon content, per cent	Endurance ratio, after reduction by drafting of				
	25 %	50 %	75 %	80 %	85 %
0.36	0.34	0.37	0.34	0.33	0.31
0.46	0.31	0.31	0.31	0.31	0.31
0.55	0.31	0.31	0.30	0.30	0.28
0.79	0.27	0.28	0.26	0.26	0.26

* If these ratios are increased by about 20 per cent so that they are comparable with other ratios which result from fatigue tests on highly polished specimens, they fall on a curve similar in slope to the curve in Fig. 128.

Peterson* commented that in conversation with him Timoshenko pointed out that, since tensile strength is based on original area, the use of the true ultimate strength, based on actual area at maximum load, or even the breaking strength, based on final area after fracture, would reduce the high endurance ratios for low-carbon steels.† The ratio between endurance limit and "cohesion strength" is more nearly constant than the usual endurance ratio.

Hatfield[303] tabulated, without comment, the endurance ratios of normalized steels of varying carbon content, the ratio being 0.55 for 0.15 per cent carbon steel, 0.37 for 0.31 and 0.60 per cent carbon steels, and only 0.28 for 0.90 per cent carbon steel; the absolute value of the endurance limit of the last-named steel was less than that of the 0.60 per cent carbon steel.

Data reported by Moore and Kommers[76] for steels containing 0.37, 0.49, 0.52, 0.53, 0.93, and 1.20 per cent carbon consistently show endurance ratios from 0.36 to 0.46 for normalized or annealed specimens and from 0.49 to 0.55 for quenched specimens tempered at temperatures as high as 650 or 760°C. (1200 or 1400°F.). Steels softened at still higher temperatures had ratios of 0.43 to 0.45.

Since the endurance limit, however secured, rises with hardness at least up to the point where increased hardness is only produced with accompanying internal stress, it is obvious that improved endurance limits may be accompanied by dangerous brittleness. McIntosh and Cockrell[164] tested steels containing about 0.10 per cent carbon, 0.05 per cent silicon, 0.50 per cent manganese, with phosphorus ranging from 0.012 to 0.085 per cent. The phosphorus raised the tensile strength from about 53,000 to about 58,000 lb. per sq. in. The endurance limit was also raised from 26,000 to 32,000 lb. per sq. in., the endurance ratio rising from 0.49 to 0.55. When testing for endurance limit on bars having a "square" notch (reduced section without fillet), the nominal endurance limits were 22,000 and 25,000 lb. per sq. in. for the low- and the high-phosphorus steels; thus the higher phosphorus steel showed no perceptibly increased susceptibility

* Private communication.

† McAdam (private communication) wrote that he had tried such ratios in 1922 but did not find them appreciably better than the ratio of endurance limit to tensile strength.

to the notch effect over that of the low-phosphorus steel. A similar trend was noted by an American Society for Testing Materials Committee[353] in "vibratory" tests on wrought iron with phosphorus ranging from 0.05 to 0.47 per cent, where the number of vibrations rose from about 9000 to about 31,000 as the phosphorus increased. Of course, the vibratory test is made at such high stresses as not to be classed as an endurance test, and the endurance limits will not be proportional to the "vibratory" figures.

The Joint Committee on Investigation of Phosphorus and Sulphur in Steel[111] studied 0.10 to 0.15 per cent carbon steel with 0.36 to 0.64 per cent manganese, 0.006 to 0.024 per cent phosphorus, 0.002 to 0.06 per cent silicon, and sulphur ranging from 0.03 to 0.18 per cent. The tensile strength varied from 44,000 to 51,000 lb. per sq. in. for annealed, and from 58,000 to 70,000 lb. per sq. in. for quenched material (varying with carbon and manganese rather than with sulphur). The endurance ratios were from 0.45 to 0.53 for annealed and from 0.46 to 0.56 for quenched material (only longitudinal specimens were used). Torsional strength and endurance limit in alternating torsion were also determined, and an approximately constant ratio was found.

It should not be concluded from these tests on very soft steels that high phosphorus or large amounts of manganese sulphide inclusions might not affect the endurance properties of harder steels.

Claims are often made for the superior fatigue resistance of certain alloy steels over carbon or other alloy steels. However, when the comparison is made on the basis of the endurance ratio, *i.e.*, on the basis of what ought to be expected from the gain in tensile strength by alloying and heat treatment, superiority on that basis alone is difficult to establish. Other desirable features, such as a high elastic ratio, greater ductility, and shock resistance, or better resistance to occasional overload, concomitant with a high endurance ratio, need to be taken into account in the actual serviceability of steel in uses which primarily call for high endurance limit but which also involve other factors.

148. Directional Differences.—Aitchison and Johnson[146] studied a normalized steel forging containing 0.29 per cent carbon, which had a tensile strength of 80,000 lb. per sq. in. both parallel

and at right angles to the direction of elongation. The endurance limit was 36,500 lb. per sq. in. for specimens cut parallel to the direction of elongation and from 30,000 to 35,000 lb. per sq. in. for specimens cut at right angles; the endurance ratio was, therefore, 0.46 for the longitudinal specimens and from 0.38 to 0.44 for the transverse specimens. When the steel was quenched and tempered to a tensile strength of 90,000 lb. per sq. in., the longitudinal endurance limit was 44,000 lb. per sq. in. and the transverse endurance limit 38,000 lb. per sq. in., the ratios being 0.49 and 0.42.

McAdam[104] also found that the endurance limits were lower on transverse than on longitudinal specimens and pointed out that these directional effects will influence the torsional endurance limit, since in torsion the stress is not applied in the direction of the grain.

Reports of longitudinal and transverse fatigue tests on hard carbon steels are very rare in the literature. Such tests were, however, reported by Gillett and Mack[154] on a nickel-chromium-molybdenum steel (quenched and tempered for a long time in order to remove internal stress) which had a tensile strength of 166,000 lb. per sq. in. and endurance limits of 89,000 and 70,000 lb. per sq. in. respectively for longitudinal and transverse directions. The endurance ratios were 0.53 and 0.42.

Thum and Buchmann[574] stated that endurance limits of transverse specimens of soft carbon steel are but 10 to 15 per cent lower than those of corresponding longitudinal specimens, while with hard carbon and alloy steels the difference is 20 to 30 per cent.

Hensel and Hengstenberg[622] found high endurance ratios, 59 to 65 per cent, on soft materials such as ingot iron and wrought iron and found no decrease in endurance ratio on transverse tests of wrought iron. The value for the transverse endurance was lower, 22,000 lb. per sq. in. as compared with 27,000 lb. per sq. in. for the longitudinal specimen, but the transverse tensile strength was 35,000 lb. per sq. in. as against 46,000 lb. per sq. in. longitudinally.

149. Significance of Endurance Data.—The endurance test evaluates only one property of the steel and does not tell whether or not other properties meet requirements; thus, a high endurance limit may be obtained by high phosphorus with accompanying

brittleness, but there is no indication of this brittleness in the endurance test. Moore has constantly emphasized this so that engineers would not overestimate the indication of an endurance test alone.

Gough[178] cited a case of normalized boiler plate compared with the same material "incorrectly heat treated." In both conditions the endurance limit was the same, 31,000 lb. per sq. in., when tested in the normal fashion, and even when a test bar with poorly rounded fillet (sharp corners) was used, the two again behaved alike, giving 18,000 lb. per sq. in. on such an endurance test. The static strengths were as follows:

Property	Normalized	Incorrectly treated
Tensile strength, lb. per sq. in. . .	62,500	60,000
Yield strength, lb. per sq. in.	43,000	41,000
Proportional limit, lb. per sq. in.	41,000	32,000
Elongation in 2 in., per cent.	42	31
Reduction of area, per cent.	63	59

The Charpy impact figures, however, were 6.3 ft-lb. for the normalized and 0.9 ft-lb. for the incorrectly treated specimen. Batson and Hyde[80] showed repeated-impact curves for "correctly and incorrectly heat-treated boiler plate" when tested with varying intensities of blows. While at the higher intensities the correctly treated one was superior, at lower intensities the correctly treated one appeared to be the weaker. However, the intensities were all so high that failure ensued after a few thousand blows. The poor condition of the incorrectly treated steel was not at all revealed by the endurance test but was shown up sharply by the single-blow impact test.

It has been often pointed out and was emphasized by Haigh[300] that ductile metals act better in actual service involving repeated stress than harder ones. Haigh related this to the position of the endurance limit with respect to the yield strength. If the endurance limit is not far below the yield strength, a slight overload will cause local yielding, relieving the stress and incidentally hardening and strengthening the metal at that point, while, if the endurance limit is far below the yield strength, an overload tends to start a crack. Kommers[312] argued that the conditions

are different when the stresses are completely reversed and there is not so much opportunity for yielding as in the case of incompletely reversed stress. Of course, the rate of reversal needs to be taken into account in the case of completely reversed stresses, because the time during which yielding in one direction can take place will have a bearing.

Aitchison(64) pointed out that proof load, endurance limit, and notched-bar impact resistance are all vital in evaluating alloys for service. Low-impact material is certainly less desirable for most uses in which overloads may occur than high-impact material, if both have the same endurance limit, and it may even be desirable to choose intentionally a material of apparently mediocre endurance limit in order to obtain a better safety factor against overload. Some engineers are coming to the conclusion that, in actual service, damage by occasional overstress is a more common cause of fatigue failure than that resulting from millions of cycles of ordinary working stress.

Moore(476) termed the property of resisting occasional overstress without developing a crack "crackless plasticity," and stated that impact tests, damping tests for evaluating the ability of the material to absorb energy, and study of properties under repeated stress after a period of overstressing have all been suggested as tending to throw light on this elusive property. French(612) discussed the last method.

The endurance limit appears to be a definite and possibly a fundamental property of steel, though its value may be slightly influenced by variations in the method of testing, just as the tensile strength is a definite property but may also be slightly affected by the test method. If the actual stress at any point, no matter how minute, exceeds the endurance limit and is often enough repeated, the steel will fail. If the steel can yield and harden locally, raising its endurance limit at the affected portion and at the same time reducing the local stress to one below the new local endurance limit, no damage is done, but, if the local overstress is maintained and often enough repeated, a sharp crack starts and progresses to failure. The nominal stress applied to a specimen in the endurance test is doubtless much below the actual local stress at stress raisers, the individual points where non-metallic inclusions are present in the interior of the metal or where scratches, notches, or sharp

changes of section are present on the exterior. If any of these stress raisers act to produce a higher local stress than is calculated, the steel responds to the *actual* stress rather than the calculated one.

C. ENDURANCE LIMIT OF CAST IRON-CARBON ALLOYS

Cast steels, cast irons, and malleable irons are now recognized as high-grade materials of construction, and for many applications their fatigue characteristics are important. Compared with wrought steels there are few data available on these three materials, too few to permit of an accurate evaluation of the fatigue characteristics.

150. Cast Steel.—According to Moore,[(191)] cast steels containing about 0.25 or 0.30 per cent carbon have endurance ratios between 40 and 50 per cent, the as-cast specimens showing the lowest ratio, and annealed, normalized, or quenched and tempered specimens usually giving higher ratios. Tapsell and Clenshaw[(240)] reported that a cast steel containing 0.53 per cent carbon and having a tensile strength of 99,000 lb. per sq. in. had an endurance limit of only 27,000 lb. per sq. in., yielding a ratio of 28 per cent.

Some interesting data on cast steel obtained by Garre and Grathoff[(521)] and reported in a recent note are listed in Table 81. The steel contained 0.18 per cent carbon, 0.25 per cent silicon, and 0.4 per cent manganese. Specimens were tested both in the as-cast condition and after annealing for 20-min. periods at several temperatures. Unfortunately, the number of cycles on which the endurance limits were based was not stated, but it is probably safe to assume that the values are based on an adequate number. The steel was cast under conditions which produced a coarse structure, and the improvement produced by annealing may exaggerate normal conditions. Annealing at temperatures of 500, 700, or 800°C. (930, 1290, or 1470°F.) did not change the ratio of torsional endurance limit to tensile strength, but annealing at 900 or 1000°C. (1650 or 1830°F.) markedly increased this ratio. The increased ratio was not accompanied by an appreciable increase in strength or hardness, but it was accompanied by a great increase in impact resistance. Annealing above the upper critical temperature thus greatly improves both the ductility and fatigue properties of a coarse-grained cast steel.

TABLE 81.—TORSIONAL ENDURANCE AND OTHER PROPERTIES OF A CAST STEEL CONTAINING 0.18 PER CENT CARBON, 0.25 PER CENT SILICON, AND 0.4 PER CENT MANGANESE*

Annealing temperature, °C.	Annealing temperature, °F.	Torsional endurance limit, lb. per sq. in.	Tensile strength, lb. per sq. in.	Ratio of torsional endurance limit to tensile strength	Yield strength, lb. per sq. in.	Elongation, per cent	Reduction of area, per cent	Impact resistance, m-kg. per sq. cm.	Brinell hardness
None		14,200	54,500	0.26	27,900	15.9	20.3	1.29	115
500	930	15,100	55,900	0.27	29,300	16.7	21.9	1.26	112
700	1290	15,400	57,600	0.27	28,800	24.6	23.8	1.33	113
800	1470	15,700	58,000	0.27	29,900	24.3	41.8	1.40	114
900	1650	19,900	61,200	0.32	36,200	22.7	42.0	4.74	117
1000	1830	21,400	62,100	0.34	36,400	25.4	50.2	5.97	121

* Garre and Grathoff.[521]

151. Cast Iron.—Eight cast irons studied by Moore and coworkers[225,226] ranged in tensile strength from 20,500 to 31,000 lb. per sq. in. and had endurance limits from 7000 to 12,000 lb. per sq. in. Their endurance ratios ranged from 33 to 46 per cent with an average of 37 per cent. Kommers[311] studied ten irons having tensile strengths from 23,000 to 51,000 lb. per sq. in. and endurance limits from 12,000 to 24,000 lb. per sq. in. Their endurance ratios ranged from 43 to 57 per cent with an average of 49 per cent and did not vary in any consistent manner with the strength of the irons; a weak iron might show a fairly high ratio and a strong iron a fairly low one.

Twenty-four irons studied by Kommers in connection with the American Society for Testing Materials' investigation of impact properties[586] gave an average endurance ratio of 50 per cent, the range being 43 to 63 per cent.

A soft iron studied by Allen, as mentioned in discussion of a paper by Bolton and Bornstein,[437] had a tensile strength of 31,500 lb. per sq. in. and an endurance ratio of 48 per cent, while a high-test iron had a tensile strength of 45,000 lb. per sq. in. and an endurance ratio of 50 per cent.

Pfannenschmidt[750] reported an endurance ratio on a molybdenum alloy cast iron of over 80 per cent, but this is evidently not a true ratio since Pfannenschmidt remarked that the tensile specimens were taken from a part of the casting which was of

greater thickness than that from which the endurance specimens were taken. He reported a ratio of 40 per cent on a chromium iron that was "too hard and brittle." After disregarding such cases, he concluded that the endurance ratio for unalloyed cast iron runs from 45 to 50 per cent and that for alloy cast iron from 55 to 65 per cent.

Moore and Picco[742] studied five lots of high-test cast iron which included a molybdenum iron, a nickel iron, and two low-silicon unalloyed irons. Of the latter, one contained 3.07 per cent carbon, 1.26 per cent silicon, 0.90 per cent manganese, 0.08 per cent sulphur, and 0.15 per cent phosphorus and, tested as cast, gave 25,000 lb. per sq. in. tensile yield strength (0.1 per cent set) and 48,000 lb. per sq. in. tensile strength, 41,700 lb. per sq. in. compressive yield strength (0.1 per cent set) and 147,000 lb. per sq. in. compressive strength with 187 Brinell. The endurance limit in completely reversed flexure was 21,000 lb. per sq. in., an endurance ratio of 44 per cent.

In this same iron, oil quenched from 870°C. (1600°F.) and tempered at 540°C. (1000°F.), the tensile yield strength and tensile strength increased to 41,800 and 76,500 lb. per sq. in. and the corresponding compressive properties to 76,000 and 183,000 lb. per sq. in. The Brinell increased to 255. The endurance limit, however, increased only to 25,000 lb. per sq. in., an endurance ratio of but 33 per cent. Heat treatment, therefore, reduced the endurance ratio from 44 to 33 per cent.

Another iron of 3.18 per cent carbon (of which 0.83 per cent was combined), 1.34 per cent silicon, 0.89 per cent manganese, 0.09 per cent sulphur, and 0.11 per cent phosphorus, which was tested only as cast, showed 255 Brinell, 33,400 and 55,000 lb. per sq. in. tensile yield and tensile strengths, respectively, and 75,500 and 159,000 lb. per sq. in. for the corresponding compressive properties. This material had a 22,000 lb. per sq. in. endurance limit, *i.e.*, an endurance ratio of 40 per cent.

The Brinell hardness does not reflect either the tensile strength or the endurance limit so closely in cast iron as it does in steel. The endurance ratios of the four irons as cast fell in the range 39 to 44 per cent. That for the heat-treated iron was decidedly lower, 33 per cent, but in view of the fact that Moore had previously found as low a ratio in a soft untreated iron, it cannot be concluded that this is characteristic of heat-treated cast irons.

Since recorded ratios for cast iron vary from 33 to 63 per cent with no indication from the ordinarily determined static properties what ratio to expect, direct experiment to establish the endurance properties of a given cast iron seems at least as imperative as with steel.

A very important property of cast iron is its low susceptibility to notches. Kommers[538] found no reduction in endurance limit in the softest iron when ordinary specimens were replaced by specimens with a square-shouldered notch, but in stronger irons with less graphite the notched specimen gave lower results. This is shown in Table 82.

TABLE 82.—EFFECT OF NOTCHES ON THE ENDURANCE LIMIT OF CAST IRONS*

Iron number	Tensile strength, lb. per sq. in.	Endurance limit		
		A—Standard specimen, lb. per sq. in.	*B*—Notched specimen, lb. per sq. in.	Decrease $\frac{A-B}{A} \times 100$, per cent
1	20,000	9,300	9,300	0
2	23,200	11,800	11,200	5.1
3	30,300	15,000	13,700	8.6
4	37,100	19,500	16,300	16.4
5	42,500	24,100	19,100	20.7

* Kommers.[538]

Kaufmann[465] studied a pearlitic iron of 3.19 per cent total carbon, 1.09 per cent silicon, 0.82 per cent manganese, 0.12 per cent phosphorus, and 0.13 per cent sulphur. It had a tensile strength of 38,000 lb. per sq. in. and an endurance limit of 14,000 lb. per sq. in.; the same endurance limit was obtained with an ordinary bar as with specimens having two types of notches. Günther,[298] who used an iron containing 3.40 per cent total carbon, 1.90 per cent silicon, 0.60 per cent manganese, 0.66 per cent phosphorus, and 0.13 per cent sulphur and having a tensile strength of 21,500 lb. per sq. in., also found a lack of susceptibility to surface notches.

Cast iron has a high damping capacity[383] (see Fig. 89, page 282), and this property is suggested by von Heydekampf[306] as a reason for its lack of susceptibility to surface notches in fatigue.

Other explanations refer to the "looseness of structure" and to the cushioning effect of the graphite.

The question of damping properties is being given much study,[511,576] especially by Föppl[699] of the Wöhler Institute in Germany. Esau and Kortum* claimed that a series of determinations of damping capacity will show the endurance limit

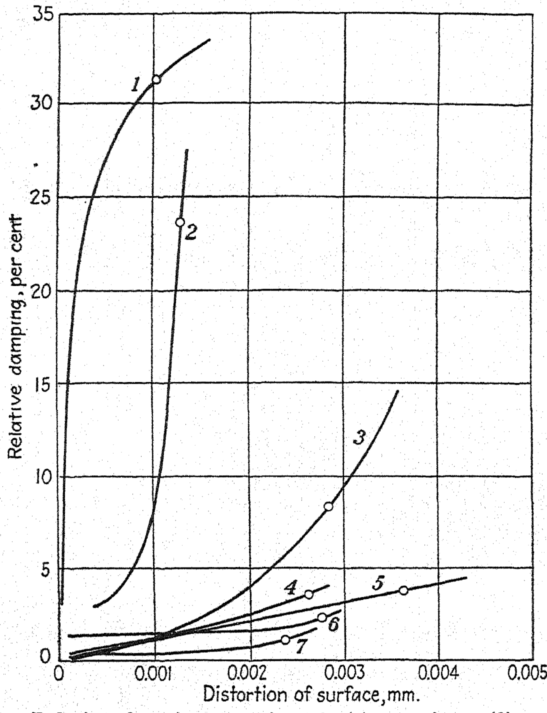

FIG. 129.—Relative damping capacity of: (1) cast iron, (2) case-hardened steel, (3) silicon steel, (4) nickel-manganese steel, (5) nickel-chromium steel, (6) chromium steel, and (7) red brass. The circles indicate the relative damping capacity corresponding to the endurance limit of the material. (*Piwowarsky*.[751])

of a metal, though this, like allegations with respect to many other accelerated methods suggested for the determination of endurance limits, needs very strict proof before acceptance.

Piwowarsky's[751] Fig. 129 indicates the relative damping of cast iron, five steels, and red brass. While, as he stated, the damping capacity may have some relation to the endurance limit, the relation, as is evident from Fig. 129, is not simple or direct.

* German Patent 568,910, Jan. 5, 1933.

It is quite probable that, as more knowledge of damping properties is obtained, some relationships between damping, "crackless plasticity," and the suitability of materials under repeated stress will become more evident. While the damping properties of iron-carbon alloys are undoubtedly properties of great importance, the state of knowledge is not sufficient to allow their evaluation at present. (See Chapter XVII for a further discussion of the damping properties of iron-carbon alloys.)

Only one set of published data on American malleable iron was mentioned in the A.S.T.M. Symposium[429] on that subject, the same data having been given previously by Schwartz.[94] According to that study, a malleable iron had an endurance of 25,000 lb. per sq. in., which is about 50 per cent of the tensile strength of this type of material. A malleable iron of 50,500 lb. per sq. in. tensile strength, tested at the Battelle Memorial Institute, had an endurance limit of 27,000 lb. per sq. in., yielding an endurance ratio of 53 per cent. Kommers* stated that three lots of malleable iron of 50,000, 55,000, and 65,000 lb. per sq. in. tensile strength each gave endurance ratios of 51 per cent. Specimens of the 50,000 tensile lot, tested with a square notch, gave an endurance limit of about 19,500 lb. per sq. in., a lowering of some 6000 lb. per sq. in., or 24 per cent, from that of the unnotched specimens. From Table 82, page 405, it will be seen that these notched and unnotched endurance limits are consistent with the trend of the data for cast irons.

D. FACTORS AFFECTING ENDURANCE LIMIT

There are a number of variables which control the life of a ferrous material in service when subjected to repeated stresses. Among these are: (1) notches or anything which leads to local stress concentration; (2) surface condition, which includes decarburization and may include small notches and other stress raisers; (3) overstressing, which may seriously damage the material; and (4) understressing. The first three usually lower the endurance limit, understressing on the contrary may actually increase it. In addition to the effect of these variables on the endurance limit as usually determined, the present chapter division also includes a brief discussion of repeated impact.

* Private communication.

152. Notches.—Surface notches tend to decrease the endurance limit of a material, as has been shown by many investigations of the endurance of specimens which were scratched, which had notches of various depths and sharpness, or which had holes drilled through them. The technique of making such tests is so different with different workers that no detailed comparison of the susceptibility of different carbon steels to reduction of fatigue resistance by such stress raisers can be made effectively.

Faulhaber, Buchholtz, and Schulz(609) found that a definite notch, or a number of tiny notches such as are present on any save a perfectly polished surface, are more detrimental upon specimens of large than upon those of small diameter; *i.e.*, there is a definite size effect when stress concentration is introduced. By taking into account, in the case of a circumferential notch, the ring formed by the diameter of the specimen, the core diameter determined by the depth of the notch, and the flanks of the notch, and calculating the moment of resistance of this ring, Faulhaber, Buchholtz, and Schulz claim to correct for the size effect and for the sharpness of the notch and thus to make it possible more accurately to determine notch sensitivity as a true constant of the material.

Characteristic results on notched specimens were given by Peterson,(556) who studied a heat-treated alloy steel of 103,500 lb. per sq. in. tensile strength and 58,000 lb. per sq. in. endurance limit (endurance ratio 56 per cent). The endurance limit of a shaft with a profiled keyway (calculated on net section) was 33,300 lb. per sq. in. A normalized 0.45 per cent carbon steel of 80,000 lb. per sq. in. tensile strength, 37,000 lb. per sq. in. endurance limit, and an endurance ratio of 46 per cent gave a notched endurance limit with this keyway of 27,400 lb. per sq. in. The percentage decrease of endurance limit was so much greater in the harder steel that the effective endurance strength of the softer steel as a keywayed shaft was only 6000 lb. per sq. in. below that of the harder steel instead of 15,000 lb. per sq. in. as for the unnotched specimens.

The curves in Fig. 130, from Mailänder,(635) show the influence of a particular notch in reducing the apparent endurance limit of steels of different strengths. As these curves show, the influence on the fatigue strength increases as the tensile strength increases. The figure also shows that austenitic steels are not

so greatly affected by a notch as other steels. The greater severity of the notch effect in steels of higher strength is also shown by Fig. 131, from a pamphlet of the Verein deutscher Ingenieure.[670]

McAdam and Clyne[739] pointed out that materials with high capacity for work hardening, like the austenitic stainless steels, are less sensitive to notch effects than those with little capacity for work hardening (such as steels already heat treated to high tensile strengths). They consider work-hardening capacity of

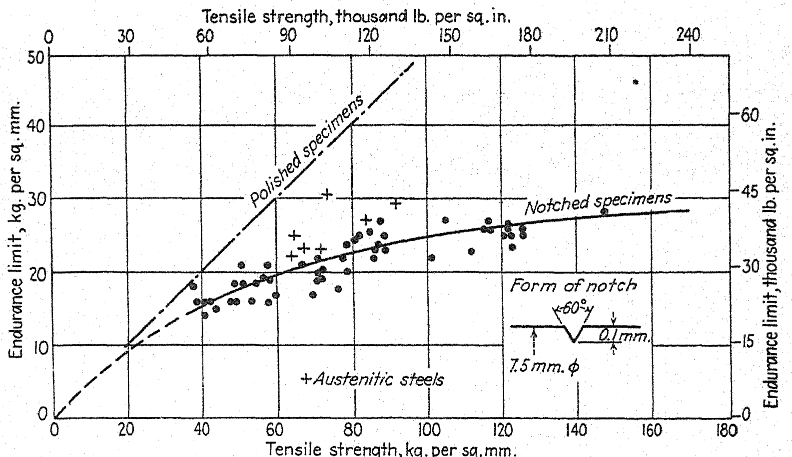

FIG. 130.—Relation between endurance limit and tensile strength for polished and notched specimens. (*Mailänder*.[635])

more direct influence upon endurance properties than damping capacity. They evaluated work-hardening capacity by means of the ratio between ordinary tensile strength and true breaking stress.

In discussing the keyway problem, Gough[178] concluded that a softer steel which can yield to local overstress without propagating a notch when actually stressed above its yield strength will yield locally, while in a harder steel with an endurance limit far below its yield strength the yield point is not reached and the crack propagates. Hankins and Ford[302] remarked that "other things being equal, the steel possessing the highest Izod value appears to be the best one to select for a spring."

153. Surface Condition.—In studying carbon and alloy spring steels, Hankins and coworkers[302,454,527] found that a spring-

tempered 0.60 per cent carbon steel, when tested as a polished specimen, had an endurance limit of 85,000 lb. per sq. in., but that, when tested as an actual spring, the endurance limit was only 22,500 lb. per sq. in. The lowering of the endurance limit was ascribed both to surface stress raisers and to the presence

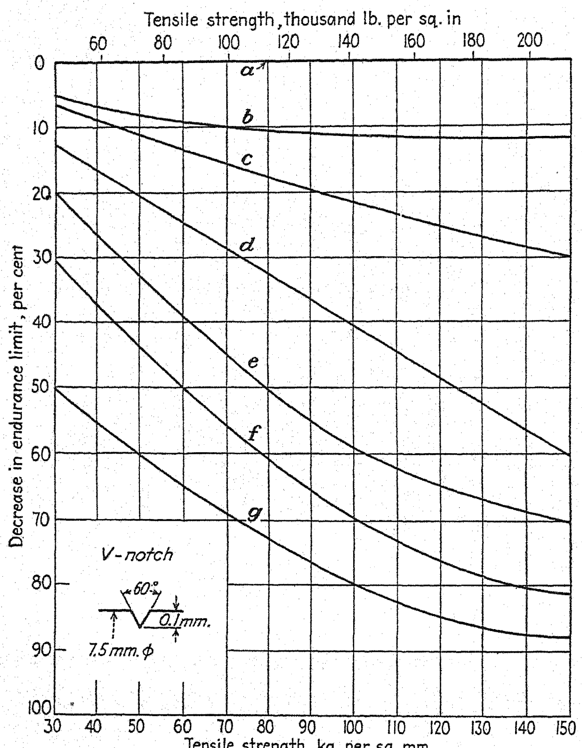

FIG. 131.—Relation between endurance limit and tensile strength for specimens having the following surfaces: *a*, polished, *b*, ground, *c*, roughened, *d*, circumferential V-notch, *e*, rolling skin, *f*, corroded in tap water, and *g*, corroded in salt water. (*V.D.I. Pamphlet.*(670))

of a soft decarburized surface layer. In rotating-beam tests or under reversed bending stresses the surface is the most highly stressed portion and, if a crack starts there, it may be propagated into the harder and stronger body of the steel. Austin(432) separated the factors of surface polish and decarburization by running tests on properly polished specimens of a steel with and without a decarburized surface layer. The steel contained

0.38 per cent carbon and 1.51 per cent manganese. In the normalized condition it had a tensile strength of 112,000 lb. per sq. in. and an endurance limit of 47,000 lb. per sq. in. when not decarburized (ratio 0.42), but with a decarburized skin the tensile strength was unchanged and the endurance limit was only 38,000 lb. per sq. in. (ratio 0.34).

As mentioned on a previous page, Gill and Goodacre[701] studied the effect of surface condition of "patented" carbon-steel

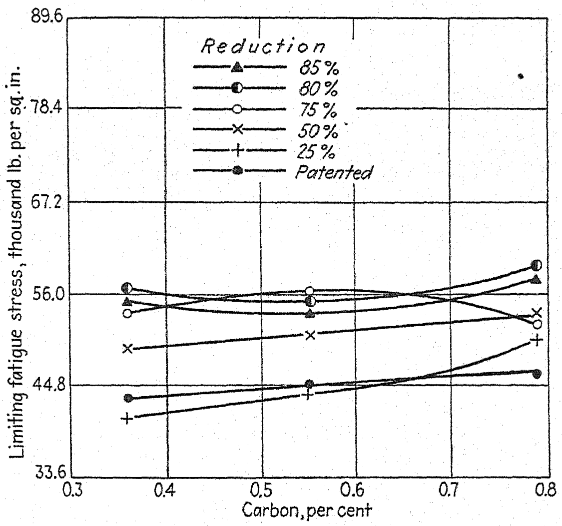

FIG. 132.—Effect of carbon content on the limiting fatigue stress of wires drawn from decarburized lead-patented rod. (*Gill and Goodacre.*[701])

wire upon endurance, using a Haigh-Robertson fatigue-testing machine. This machine rotates a wire which is bowed out by end thrust to produce higher stress at the mid-length than at the grips and thus allows testing the wire without giving it a reduced section. The machine operates at the high speed of 14,000 cycles per min. They found that the endurance limit of medium- and high-carbon wires drawn from patented rods which had a thin decarburized skin—0.002 in. thick—was so greatly affected by this decarburized surface that the effect of carbon content and of amount of drafting was almost completely masked. This is evident by comparing Fig. 132, the endurance limits of the decarburized specimens, with Fig. 133, which shows the endur-

ance limits of wire of the same composition, drawn with the same reductions but from rod from which the decarburized skin had been removed by machining. In the latter the endurance limits reflect clearly the effect of carbon content and of amount of drafting. As shown in Figs. 133 and 134, the endurance limit increases with the amount of cold working, but, as noted on page 391, the endurance ratio changes very slightly. As is also evident from Fig. 133, the limiting fatigue stress increases with the carbon content for all degrees of cold working used by Gill and

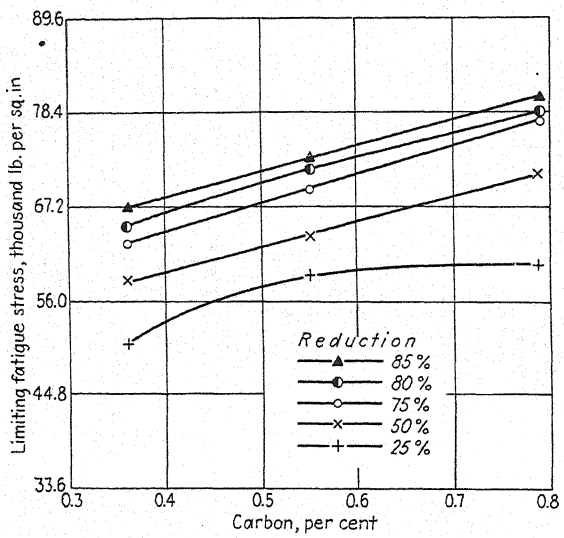

FIG. 133.—Same as Fig. 132, but rods were not decarburized. *(Gill and Goodacre.*[701]*)*

Goodacre, but, as the tensile strength is affected to a somewhat greater degree, the ratio as shown on page 393 is lower for the high-carbon steels. All of the steels tested had a relatively low ratio, varying from 26 to about 37 per cent.

Gill and Goodacre determined the limiting fatigue stress on wire with the surface in the condition which it had when it came from the dies (both with and without a decarburized skin). These investigators showed, in the case of 0.79 per cent carbon steel, that polishing the surface with 000 emery to remove die marks and other slight imperfections raised the endurance limit some 25 per cent. As quoted on page 393, a specimen

containing 0.79 per cent carbon, free from decarburization, had a ratio of 0.24 as it came from the dies; after polishing, the ratio was 0.30.

The possibility of correlating fatigue strength and overdrawing in drafting is also mentioned on page 392. There was evidence of overdrawing in the 0.79 per cent carbon wire which had a decarburized skin and which was reduced 75 per cent or more in drafting but none in the same wire drawn from rod which was not decarburized. An important feature of this investigation

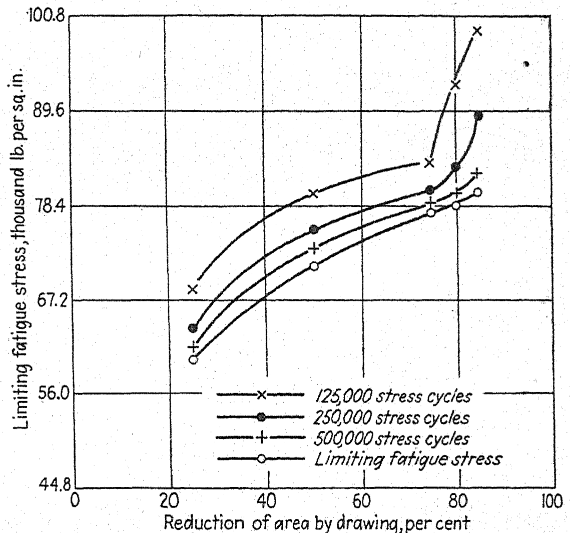

FIG. 134.—Endurance curves of 0.79 per cent carbon patented wire, drawn from rod free from decarburization. (*Gill and Goodacre.*[701])

is the attention paid to results at stresses, above the endurance limit, that cause relatively early failure. While it is beyond the scope of this monograph to discuss all the high-stress results of Gill and Goodacre on specimens with decarburized surfaces, Fig. 134 gives some interesting data on the way the stress for 125,000 cycles life increases above 75 per cent reduction.

High-carbon steel wire (0.77 per cent carbon), quenched and tempered and galvanized, was studied by Shelton,[491] who used an ingenious method. With the galvanized coating on the wire the tensile strength was between 215,000 and 230,000 lb. per sq. in. and the endurance limit was only 51,000 lb. per sq. in., an

endurance ratio of 23 per cent. Removing the brittle zinc layer by pickling, but leaving a rough surface, raised the tensile strength to 228,000 lb. per sq. in. and the endurance limit to 59,000 lb. per sq. in., a ratio of 28 per cent, while polishing the surface raised the endurance limit to over 100,000 lb. per sq. in., a ratio of over 44 per cent. The sensitivity to surface cracks, notches, and roughness is thus clearly shown. Additional data, including axial-loading tests at various stress ranges, have been more recently reported by Shelton and Swanger.(816)

Under conditions of repeated stress where the maximum stress is applied at the surface, a strong, high-carbon skin, as in case-carburized steel, greatly raises the endurance limit over that of the core. Woodvine(143) and later McMullan(808) discussed the stress distribution in a duplex material and the relations between endurance limits in case and core which will bring the nucleus of failure in a rotating-beam specimen of carburized steel below the surface rather than at the surface. Moore(224) showed that by deeply carburizing iron (the case extending one-fourth the diameter of the test specimen), the endurance limit was raised from 26,000 to 68,000 lb. per sq. in., but he commented that this increase would not be so marked under axial loading. Hensel and Hengstenberg(622) made axial-loading tests on very deeply carburized ingot-iron specimens and found the endurance limit to rise only from 28,000 to 45,000 lb. per sq. in. About the same ratio of increase was met in deeply carburized wrought iron, the endurance limit rising from 26,500 to 40,500 lb. per sq. in. These authors made the interesting observation that, although the inclusions in soft wrought iron did not have much effect upon the endurance limit, when the material was carburized failure occurred at the fillets, *i.e.*, at locations of stress concentration, though these were of the proper radius to prevent fracture there in materials which are not brittle. Stress raisers or inclusions thus have a much worse effect in very hard iron-carbon alloys than in soft ductile materials.

A nitrided skin has a strengthening effect in fatigue which is similar to that of a carburized case. Harder(455) collected information on this subject. As the nitriding steels are usually alloyed steels, they are not discussed here.

Another method of producing a fatigue-resistant surface layer that will be effective in rotating-beam tests or in stressing by

repeated bending is cold working the surface. Along with other alloys, Döring[365] used a 0.35 per cent carbon steel for testing after rolling a cylindrical specimen under pressure of three rollers. Such cold working of the surface of this steel raised the endurance limit of the specimens from 8 to 12 per cent. Several articles[460,591,605,787] giving detailed results of the behavior of specimens whose surfaces are cold worked in various ways have recently appeared in the reports of the Wöhler Institute.

Zander[351] showed that a chisel mark on an endurance specimen is not so severe in its action in lowering the endurance limit of a soft steel as is a notch of similar shape and depth machined into the surface, which may be attributed to the beneficial cold working of the material at the base of the chisel mark.

154. Overstressing.—Cold work that distorts the material increases hardness and endurance limit. However, specimens on the high-stress end of the *S-N* curve, *i.e.*, run at an overstress but at an alternately reversed one, are not measurably distorted at normal speeds of endurance testing. They may, nevertheless, show a hardening analogous to that of cold working. Moore and Ver[397] found increased hardness of Armco iron and 0.20 per cent carbon steel in unfractured but overstressed metal in rotating-beam tests. Such specimens, tested in tension, showed markedly lower reduction of area than specimens not overstressed. In this work the rotating-beam specimens had two reduced-diameter necks, and the tests were made on the unbroken neck after fracture had occurred at the other. The hardened material was, therefore, classed as "fatigue damaged."

The fact that repeated overstress at sufficient intensity and for a sufficient number of cycles causes damage (and, as is pointed out later, that understressing causes improvement) is utilized in fatigue testing to ascertain whether the endurance limit has actually been reached after a large number of cycles. After running at one stress, the stress upon the specimen is increased, and the life of the specimen at the second, higher, stress compared with that of a second specimen of the virgin material initially loaded at this higher stress. If the life is materially shorter than that of the second specimen, "fatigue damage" is shown at the initial stress, and that initial stress is taken as being *above* the true endurance limit for virgin material. If, on the other hand, the life at the higher stress is materially above that

of the second specimen, improvement of the material at the initial stress is shown, and this stress is taken as *not* being above the endurance limit. Johnson and Oberg[310] used this method to good effect on duralumin, and it has been much used on steel;[119,295,367] its application also was discussed in A.S.T.M. reports.[284]

Moore and Wishart[642] tentatively suggested that a series of test bars be run, each at different stresses for about 1.5 million cycles, then removed and tested in tension, when material improved by understressing will show slightly higher tensile strength and "fatigue-damaged" material a slightly lower one.

Nevertheless, it is possible for a specimen to withstand a limited number of repetitions of overstress without its life being lowered when tested further at the endurance limit of a virgin specimen. Moore and Kommers[224] stated:

A stress which is a considerable percentage above the endurance limit may be applied from 1000 to 5000 times without greatly influencing the fatigue strength under subsequent application of lower stresses. Results on a heat-treated 0.49 per cent carbon steel in the sorbitic condition showed that a stress 10 and 20 per cent above the endurance limit applied 5000 times, a 29 per cent overstress applied 1000 times, and a 38 per cent overstress applied 100 times did not appreciably reduce the endurance limit as subsequently determined. However, an overstress of 35 per cent applied 1000 times reduced the endurance limit 4 per cent, while an overstress of 29 per cent applied 5000 times reduced the endurance limit about 11 per cent.

A heat-treated 1.20 per cent carbon steel in the sorbitic condition whose original endurance limit was 50,000 lb. per sq. in. was subjected to 20 per cent overstress for 5000 and 10,000 cycles, respectively.

The endurance was reduced 12 per cent and 14 per cent respectively. Comparing the result of 20 per cent overstress applied 5000 times in the case of the 0.49 per cent carbon steel, whose Brinell hardness was 197, and the 1.20 per cent carbon steel, whose Brinell hardness was 369, it is seen that the harder steel was much more influenced by the overstress than the softer steel. It should be noted here that the absolute value of the overstress was practically the same in the two cases, because the endurance limits did not differ greatly.

A definite number of cycles of a certain degree of overstress can thus be withstood without lowering the endurance limit. Gough[178] said:

It is apparent that for any degree of overstress there is a limiting number of repetitions below which the fatigue range of the material is not affected; further, that this limiting number decreases as the amount of overstress increases. The influence of the overstress is also obviously greater on the hard than on the softer steel, a result which would be expected.

It was shown by French(612) that in some rare cases, perhaps ascribable to a precipitation-hardening effect induced by stress, a preliminary overstress may even raise the endurance limit, though, of course, too high a stress or too many repetitions of it cause damage.

As French pointed out, there is a "damage line" positioned below the high-stress portion of the *S-N* curve and conforming more or less to its shape (see Fig. 127). Hence, the location of the high-stress branch of the *S-N* curve, which is often not accurately determined because too few specimens are used, may be useful in evaluating the ability of a given steel to withstand overstress.

The matter is complicated by the fact that, at such high stresses that the proportional limit is materially exceeded, the calculated stress is only a nominal one(224) and the static stress-strain diagram has to be taken into consideration. Unfortunately, the shape of the curve is somewhat dependent upon the test conditions, *i.e.*, shape of specimen, manner of loading, etc., so that evaluation of the behavior of steels of different carbon content and different grain size and structure by this method cannot yet be made conclusively on the basis of published data. Uniform test methods will be required in order to eliminate variables other than those to be studied.

155. Understressing.—The fact that repeated stressing below the endurance limit may improve the material so that when retested at a higher stress it lasts far longer than a virgin bar would, had it initially been tested only at the higher stress, has been observed by many workers(103,104,178,224,284,472) in fatigue of metals.

While the material is not measurably distorted, it is hardened and strengthened. Up to 30 per cent increase in endurance limit has been secured in this manner by Gough,(178) and Moore(224) has noted over 40 per cent increase. The hardness and tensile strength are also increased.

The phenomenon of strengthening by understressing appears to be a general one in steel, though it is not clear just how closely the endurance limit must be approached or how many cycles must be applied to secure a given degree of strengthening. Most observers(178,384) seem to agree that ingot iron is hardly susceptible to strengthening by this means, since only 3 to 5 per cent improvement has been obtained by some, while Kommers(384) found only a 10 per cent improvement on understressing ingot iron.

Swanger and France(572) noted that quenched and tempered steels were strengthened by understressing but that annealed carbon steels were not. However, Moore(224) found distinct strengthening in an annealed 0.49 per cent carbon steel, as well as in several normalized steels, and Freeman, Dowdell, and Berry(210) found it well marked in rail steel. Although ingot iron showed little effect, Moore and Jasper(132) found that wrought iron was perceptibly strengthened. Moore, Lyon, and Inglis(226) have reported strengthening in cast iron, and Kommers(384) found that soft cast iron of 20,000 lb. per sq. in. tensile strength and a virgin endurance limit of 9300 lb. per sq. in., stressed at 9000 lb. per sq. in. for 1.5 million cycles or more (up to 40 million), had a new endurance limit of 11,600 lb. per sq. in. or a 25 per cent improvement. Stressing for 20 million cycles at 9300 lb. per sq. in. raised it 31 per cent to 12,300 lb. per sq. in., while 20 million cycles at 6500 to 8000 lb. per sq. in. raised it only 16 per cent to 10,800 lb. per sq. in. He has also* found malleable iron of 50,000 lb. per sq. in. tensile strength capable of 10 to 15 per cent improvement by understressing.

It is obvious that, because the properties are affected by understressing in the early part of the test,(178) endurance limits cannot be correctly determined by a "step-up" method, *i.e.*, by running a single bar at a low stress and raising the stress by steps till it breaks.

156. Repeated Impact.—Resistance of a material to repeated impact does not necessarily vary directly with the endurance limit; the test as ordinarily carried out does not appear to measure a simple and fundamental property of a material.

One might expect that a repeated-impact test on notched bars might show the service-endurance properties, combining

* Private communication.

the effect of the magnitude of the endurance limit and the propensity toward notch propagation. Mathews, quoted by Gillett and Mack,(154) expressed the opinion that the repeated-impact test does measure ability to stand overload. Most repeated-impact tests have been made at such high intensities of blow that failure ensued after a few thousand blows. According to Aitchison(64) and McAdam(104) different steels can be rated by the results of repeated-impact tests in approximately the same order as determined by single-blow notched-bar tests, provided the stresses for the former produce failure after a few hundred or thousand blows. For lower stresses which produce fracture after 100,000 or more blows, the order in which steels may be rated by repeated impact is the same as when rated by fatigue tests. For intermediate stresses which produce failure by repeated impact after more than a few thousand but less than 100,000 blows, the data cannot be readily interpreted. Owing to the lack of data on different steels over a wide enough range of intensities of blow, evaluation of the effect of carbon in steel by means of the repeated-impact test cannot yet be made effectively.

E. CORROSION FATIGUE AND PROTECTIVE COATINGS

If iron or steel is corroded and then subjected to repeated stress, the corrosion pits or roughness on the surface lower its endurance limit. According to McAdam and Clyne,(739) a correlation of considerable data on the influence of prior stressless corrosion on fatigue has shown that the results are applicable to a study of the influence of notches on the fatigue of metals. The effect of such "chemically formed notches" is usually "due less to general loss of section than to stress concentration caused by corrosion pits."

If corrosion occurs *simultaneously* with repeated stress (called stress corrosion or corrosion fatigue by McAdam and Clyne), as when a stream of water plays upon an endurance-test specimen, the effect is much more marked.* Under such conditions the strongest heat-treated or cold-worked, but corrodible, steel

* Gough and Sopwith(525) allege that this is not the case, on the basis of their study of a single crystal of aluminum which did not pit under corrosive conditions. This single observation, however, can hardly be taken as overthrowing the generally accepted belief.

behaves almost as poorly as the weakest, annealed, but equally corrodible, material.

157. Corrosion Fatigue.—Steels which have an endurance limit of as much as 100,000 lb. per sq. in. when not under corrosive influences may, when under corrosive conditions, show a value

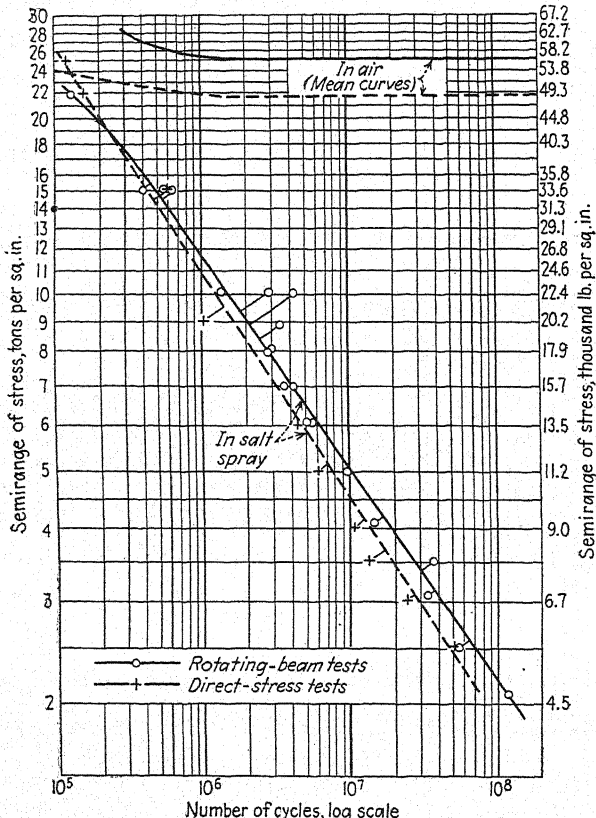

FIG. 135.—Stress-cycle curves for a 0.5 per cent carbon steel, in air and in salt spray. (*Gough and Sopwith.*[615])

of only 20,000 ± 5000 lb. per sq. in. Furthermore, for corrodible steels, there is probably no definite stress at which they will indefinitely resist a repeated stress, *i.e.*, they may have, under some corrosive conditions, no true "corrosion-fatigue limit." The stress-cycle curves shown in Fig. 135, from Gough and Sopwith,[615] for a carbon steel tested in a salt spray show that even after 100 million cycles no endurance limit had been found.

Since the corrodibility of iron and the iron-carbon alloys—steels, cast and malleable irons—while distinguishable, is of the same order of magnitude (in comparison with the more corrosion-resistant alloys such as those of iron with sufficiently high chromium contents or of copper with a variety of alloying elements), differentiation of the iron-carbon alloys according to resistance to corrosion fatigue is a difficult matter. Much more depends on the corrosive environment than upon variations

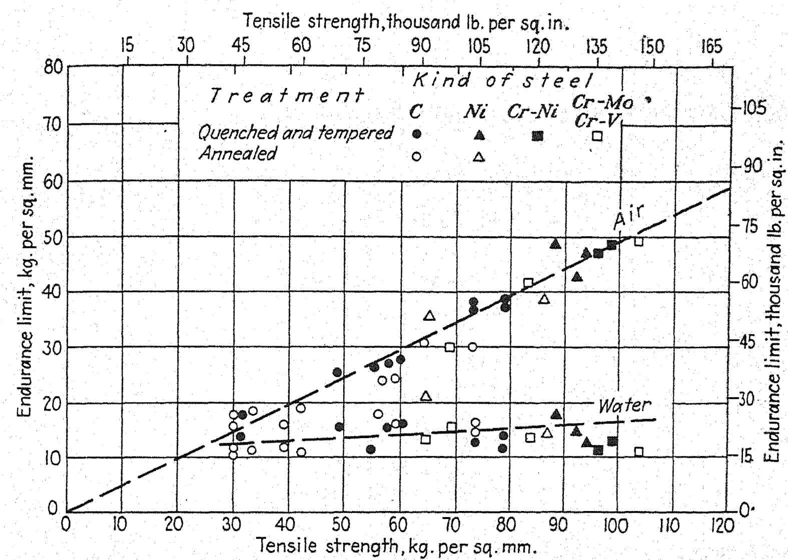

FIG. 136.—Endurance limits of carbon and alloy steels tested in air and in water. (*McAdam's data as interpreted by Mailänder.*(635))

in the composition and structure of the iron-carbon alloys. Hence, the interested reader is referred for specific details to a large number of papers by McAdam,* to an extended summary by Gough,(523) to a briefer one by Moore,(641) and to a recent résumé by McAdam and Clyne.(739)

Suffice it to remark here that over the range of compositions of plain carbon steels from ingot iron to 1.10 per cent carbon, whether quenched and tempered or annealed, neither increase in carbon content nor improvement in tensile strength due to heat treatment is efficacious in improving performance under com-

* See bibliography, reference Nos. 189, 220, 221, 222, 269, 270, 322, 323, 324, 388, 389, 390, 472, 473, and 548.

bined corrosion and repeated stress. This is conveniently shown in Fig. 136, plotted by Mailänder[635] from McAdam's data.

Obviously, it is the corrosion resistance rather than the resistance to repeated stress in air that counts most heavily. In other words, the skin of corrosion products must be flexible enough to stand repeated stress without rupture or failure to keep the corroding medium away from the metal at every point. Plain carbon and low-alloy steels do not have such skins, and only such films as those of the type formed upon the high-chromium steels are effective in resisting cyclic stress without cracking.

During the period in which corrosion and repeated stress are acting, within the life of a corrodible steel, the time during which corrosion acts is as important as the number of stress cycles applied. Hence the higher the rate of repetition of stress the higher the apparent corrosion-fatigue limit—the lower the rate, the lower the corrosion-fatigue limit. Therefore no single stress can be set as this limit without stating all test conditions. The interrelations among the variables, as worked out by McAdam, are complex.

Since nothing much can be done to combat corrosion fatigue by alloying—except using steels containing high chromium—or by heat treatment, engineering attention has to be focused, when the ordinary carbon and low-alloy steels are used under repeated stress under corrosive conditions, upon preventing corrosion by some film or coating.

158. Protective Coatings.—Inhibitors, such as chromates, were studied by Speller, McCorkle, and Mumma[344] and by Gould,[703] who found that a limited degree of protection can be afforded where it is possible thus to alter the composition of the corroding medium. Under most corrosive service conditions too much inhibitor is required to make the method practical.

A more generally applicable method is the use of metallic protective coatings; obviously coatings that are cathodic to the base steel will only be effective when the coating is perfect and continuous. The relatively small effect of such a metal coating when not absolutely perfect is shown by Honeger's tests, quoted by Gough.[523] A steel of 43,000 lb. per sq. in. endurance limit in air gave the same after chromium plating. In a water stream the unplated steel had an endurance value of about 20,000 lb.

per sq. in. on the basis of 200 million cycles, and the chromium-plated specimen a value of 25,000. According to Cleaves and Thompson,[794] chromium, although anodic in the usual order of the electrochemical series, behaves as a neutral coating which neither inhibits nor accelerates corrosion electrochemically. On the other hand, tests by McAdam[324] with heat-treated chromium-vanadium spring steel indicated that a cadmium plating, 0.0002 in. thick, more than doubled the fatigue limit under corrosion by fresh tap water. Cleaves and Thompson[794] stated that, although cadmium is cathodic to iron in the usual electrochemical series, it may be anodic in tap water.

Some plated steels apparently have lower endurance limits in air, without corrosion, than the same steels unplated. Lea[316] and Inglis and Lake[534] found that nickel plating lowered the endurance limit of a 0.14 per cent carbon steel about a third and suggested that this may be due to the propagation of a crack from the coating through the steel itself. The question of notch propagation has been summarized[296] from this point of view. Barklie and Davies[355] discussed the problem from the point of view of stresses in the electrodeposited coatings. From the point of view of prevention of corrosion fatigue, anodic coatings of zinc or cadmium (cadmium, as stated above, being probably anodic to iron in sufficiently dilute solutions of some electrolytes) have had most attention.

Shelton's data on air fatigue of steel with hot-galvanized coatings have already been cited (page 413). Haigh[299] and Gough[523] applied hot-galvanized coatings to paravane towing cables which were failing in service in sea water with a resulting vast improvement in life, *i.e.*, in resistance to corrosion fatigue, although, as Swanger and France[572] have shown, the air endurance of hard steels is reduced by hot galvanizing. Electroplated zinc or cadmium coatings upon initially smooth steel surfaces were found, according to Swanger and France, not to reduce the air-endurance limit. Harvey[374,528] has indicated that, where pickling embrittled or roughened the surface, the air endurance fell from that cause; but he has also shown that, if that can be avoided, the plated steel should have about the same endurance under corrosion fatigue as in air fatigue as long as the plating remains sufficiently complete and impervious. Harvey ascribed the protection of cadmium to the impervious-

ness and flexibility of the skin of corrosion products rather than to electrochemical protection by the metallic cadmium.

Temporary embrittlement by hydrogen absorbed in plating was discussed by Lea,[316] and Vollmer and Westcott[419] found that plain uncoated carbon steels became temporarily embrittled by moist hydrogen sulphide, with disastrous results under repeated stress. Prevention of access of hydrogen sulphide to the steel by galvanizing was effective. In this case a copper coating was also found useful.

Gough and Sopwith[524,799] reported some tests in vacuum to see whether "air corrosion" was a factor in fatigue. While they reported some such effect on an annealed 70:30 brass, which observation needs corroboration by other observers before acceptance, the two steels tested showed no detectable difference between air and vacuum.

F. AUTHOR'S SUMMARY

1. Iron and steel have a fatigue or endurance limit which is a definite property of the material in the same sense as the tensile strength is a definite property. This limit represents the maximum stress that can be indefinitely applied as alternate compression and tension to the material without producing failure. This stress may or may not be above the proportional limit of the material.

2. The endurance limit is obtained by subjecting different samples of the material to reversed stresses and determining the number of cycles of the reversed stress necessary to produce failure. If the number of reversals, N, required to break the specimen is plotted against the stress, S, and a smooth curve is drawn through the points, the endurance limit is the stress at which the curve becomes parallel to the N axis.

3. The rotating-beam test is most frequently used for determining the endurance, but axial loading is sometimes used. Endurance limits accurately obtained by the latter type of test are not greater than those accurately obtained by the former. Differences between results obtained by the two types of tests may be due primarily to inhomogeneity of the material or to the effect of small discontinuities such as scratches and inclusions; the influence of these factors is greater upon the axially loaded specimen because of the greater volume stressed.

4. For sound wrought ferrous alloys the endurance limit should be roughly half of the tensile strength. The ratio of endurance limit to tensile strength is called the "endurance ratio." While as a first approximation it may be assumed that the endurance ratio is 50 per cent, the ratio is *not* sufficiently constant to make it unnecessary actually to determine endurance limits. Very hard steels have a low endurance ratio, probably due to the existence of internal stresses. Naturally, the fatigue limit is also roughly proportional to the Brinell hardness, because the Brinell hardness and tensile strength are proportional, but it bears no certain relationship to any other property.

5. Variations in composition or treatment which influence hardness and strength also influence the endurance limit. Cold work increases endurance limit as well as strength. A material may have a high endurance limit and yet be dangerously brittle, and endurance limit alone cannot be used for determining the suitability of a material for certain commercial applications. Cold working to such an extreme that incipient internal cracks are formed or high internal stresses introduced lowers the endurance limit and endurance ratio. The effect may not be marked upon the static properties; thus endurance testing may be a means of detecting too drastic cold working.

6. Steels in the as-cast condition generally have a lower endurance ratio than heat-treated cast steels. Endurance ratios of from 37 to 50 per cent have been reported for cast iron. Cast iron is less susceptible than steel to stress raisers such as scratches or notches. High-strength cast irons appear to be somewhat more affected by notches than the low-strength irons.

7. Quenched and tempered steels tend to have higher endurance ratios than annealed or normalized steels. The endurance ratio of quenched steels and of those tempered at low temperatures is usually low, rising with higher tempering temperatures and then falling as the fully softened condition is approached. Other conditions being similar, in normalized and annealed steels there is a tendency for the endurance ratio to fall as the carbon increases.

8. A decarburized skin greatly lowers the endurance limit. Overstressing may, and usually does, seriously damage the material, though some materials will stand some overstressing and, to a degree, may be benefited by it. Understressing may,

and in most alloys does, actually raise the endurance limit. Materials vary in their susceptibility to damage by overstress and to improvement by understress.

9. Repeated-impact tests yield values which may be correlated with endurance limit when failure occurs after a great number of blows, and values which may be correlated with single-blow impact result when failure is brought about by a small number of blows.

10. There is probably no true corrosion-fatigue limit. Corrodible ferrous alloys do not exhibit great differences in resistance to corrosion fatigue.

11. Highly corrosion-resistant steels, such as those of high chromium content, are much superior to the plain carbon steels under repeated stress and simultaneous corrosion. Zinc or cadmium coatings are useful upon carbon steels subject to simultaneous corrosion and repeated stress as long as they remain continuous. If the surface of the steel is roughened in the process of applying the coatings, the stress raisers so produced may adversely affect the endurance. Inhibitors which prodvce protective films (passive surfaces) on steel are effective to some degree, but many corrosive conditions require too high a concentration of inhibitor to make this method of protection commercially useful.

12. The endurance properties of metals and alloys are detrimentally affected by notches, keyways, or anything that leads to local stress concentration. The harder steels are more adversely affected by notches than the softer ones, so much so that with severe notches there may be no great difference between the notched endurance strength of two steels which in the unnotched condition show great differences. Most fatigue failures in actual service occur in parts subject to stress concentration and to occasional overstress. A combination of toughness, such as is evidenced by high resistance to the single-blow notched-bar impact test, and high endurance limit best serves to combat such conditions.

CHAPTER XII

EFFECT OF TEMPERATURE ON PROPERTIES: I. SHORT-TIME TESTS

General Trend of Static Properties—General Trend of Dynamic Properties—Properties at Temperatures below the Blue-heat Range—The Blue-heat Range and the Range of Secondary Brittleness—Cast Ferrous Alloys—Author's Summary

Little information is available on the effect of temperature on the properties of "pure" iron-carbon alloys, but a very extensive literature on the properties of commercial steels at high and low temperatures is listed in bibliographies and summarized in symposia, reports, books, and comprehensive articles.* This literature contains upward of 1500 references. Notwithstanding the fact that a large amount of work has been done and the data are excellently summarized, many fundamental facts about high- and low-temperature properties of carbon steels cannot yet be stated definitely.

The technique of testing at other than room temperature is difficult, the equipment expensive, and the work, especially elevated-temperature creep testing, very tedious. Carbon steels lose strength so markedly at high temperatures that alloy steels and special heat-resisting alloys are universally used where service conditions are unusually severe; the attention of investigators, therefore, has been largely given to alloy steels rather than carbon steels.

High-temperature properties of metals are determined by two principal methods: (1) short-time tests which, as is self-evident, are made by heating the specimens to the desired temperature, holding at this temperature until thermal equilibrium is attained, and then testing; and (2) long-time, or creep, tests in which the time factor and the stress which causes failure at the given temperature are considered to be of equal, or almost equal, importance. Although a large part of the attention paid to

* See bibliography, reference Nos. 22, 67, 78, 110, 153, 187, 204, 212, 244, 260, 328, 368, 380, 430, 431, 452, 496, 560, and 716.

high-temperature testing has, in the last few years, been on the determination of creep, both short- and long-time tests have their advantages in determining the fitness of a material for industrial use, and both have received attention in this monograph; the former—short-time tests—in the present chapter, and the latter in the next chapter.

The effect of subnormal temperatures on ferrous alloys is determined by short-time tests, usually by impact. The properties so determined are of importance in evaluating the worth of a steel to be used in the construction of refrigerating machinery, aircraft which may operate at high altitudes, and structural steels used in very cold climates.

A. GENERAL TREND OF STATIC PROPERTIES

As a general rule, the hardness and strength of a metallic material vary inversely with the temperature, while the ductility varies directly. However, not only does the rate of change of properties vary with temperature, but in some alloys actual maxima or minima are observed in curves representing a property as a function of temperature. For example, many steels have a lower strength at room temperature than at somewhat higher or lower temperatures. In order really to determine just how the properties of a steel vary with temperature, it is not only necessary to test the material over a wide range of temperatures, but tests must be made at rather small temperature intervals.

The data discussed in the first part of this chapter were obtained on wrought carbon steels; the elevated-temperature properties of cast materials are discussed at the end of the chapter (pages 479 to 491).

159. General Relation of Tensile Properties to Temperature.—A picture of the change in strength of steels with varying carbon content, over a wide range of temperatures, is given in the early report of Robin[18] (see Fig. 137) in which crushing tests are described which were made at temperatures ranging from that of liquid air to at least 1100°C. (2010°F.). His data on strength at hot-working temperatures may be supplemented by those of Ellis,[519] shown in Fig. 138, who also gave curves (not here shown) for five steels containing 0.90 per cent carbon, which closely follow the curve in Fig. 138 for the steel containing 0.85 per cent carbon. These data show that the increased strength

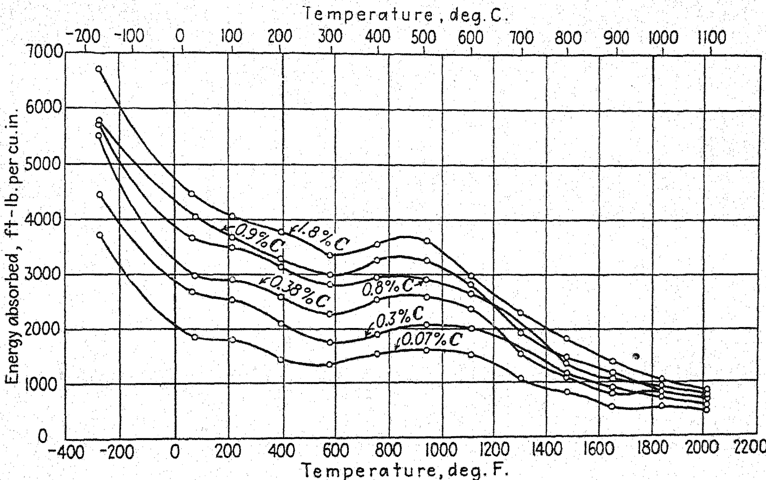

FIG. 137.—Effect of temperature on the impact energy absorbed in crushing untreated carbon steels. (*Robin.*[18])

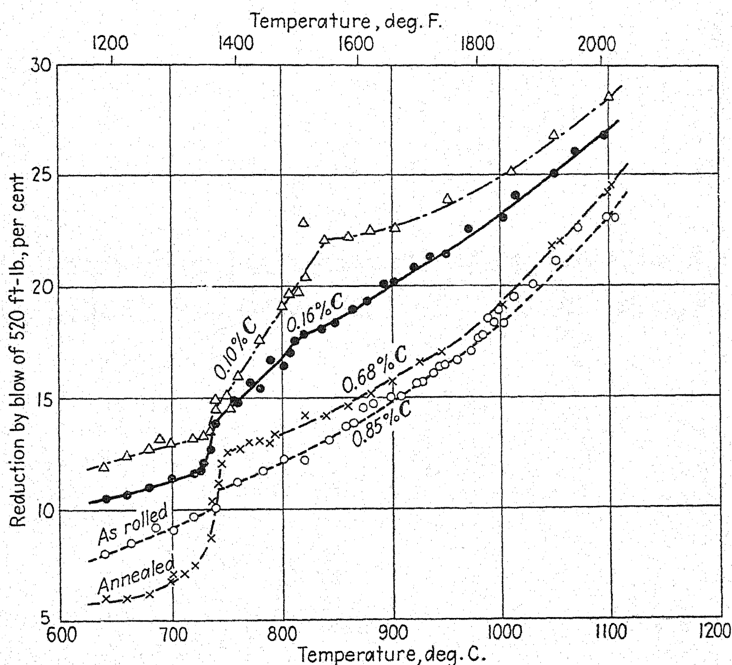

FIG. 138.—Effect of temperature on the amount of reduction by forging for four carbon steels. (*Ellis.*[519])

imparted by an increase in carbon content is retained to some degree at all temperatures, even in the austenitic range.

Grün[707] made short-time tensile tests at 400, 500, and 600°C. (750, 930, and 1110°F.) on steels with carbon from 0.05 to 0.40 per cent, which indicate that at the higher temperatures little if any strength is added by increase in carbon from 0.30 to 0.40 per cent. He also applied at 400 and 500°C. (750 and 930°F.) so-called accelerated creep tests of a type which, as is discussed later, is not generally accepted as giving reliable information

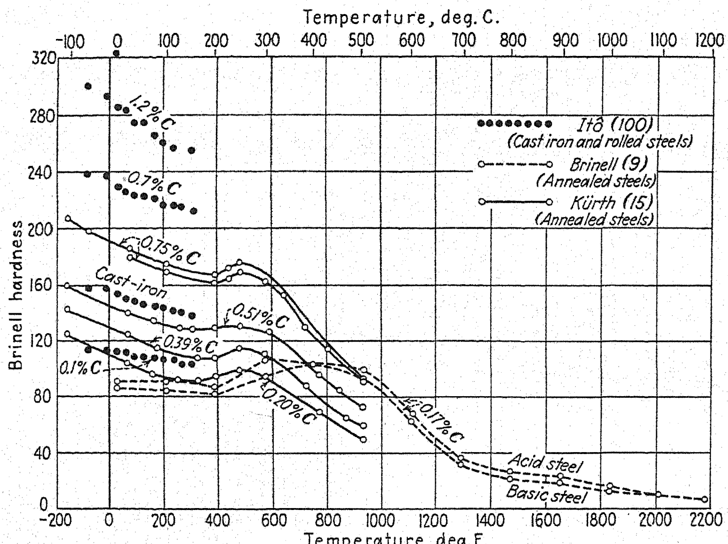

FIG. 139.—Effect of temperature on the Brinell hardness of several carbon steels and a cast iron.

on the long-time load-carrying ability but which can be considered as short-time tensile tests made more slowly than usual. These tests indicate that even within a few hours the difference in apparent strength conferred by the carbon content is largely lost, *i.e.*, the higher carbon steels show a much smaller degree of superiority.

Compression tests made by Petrenko[135] at the National Bureau of Standards on wrought iron and steels of 0.10, 0.32, and 0.95 per cent carbon at temperatures ranging from room temperature to 1100°C. (2010°F.) indicate similar tendencies. With these may be compared later compression tests by Sale,[763]

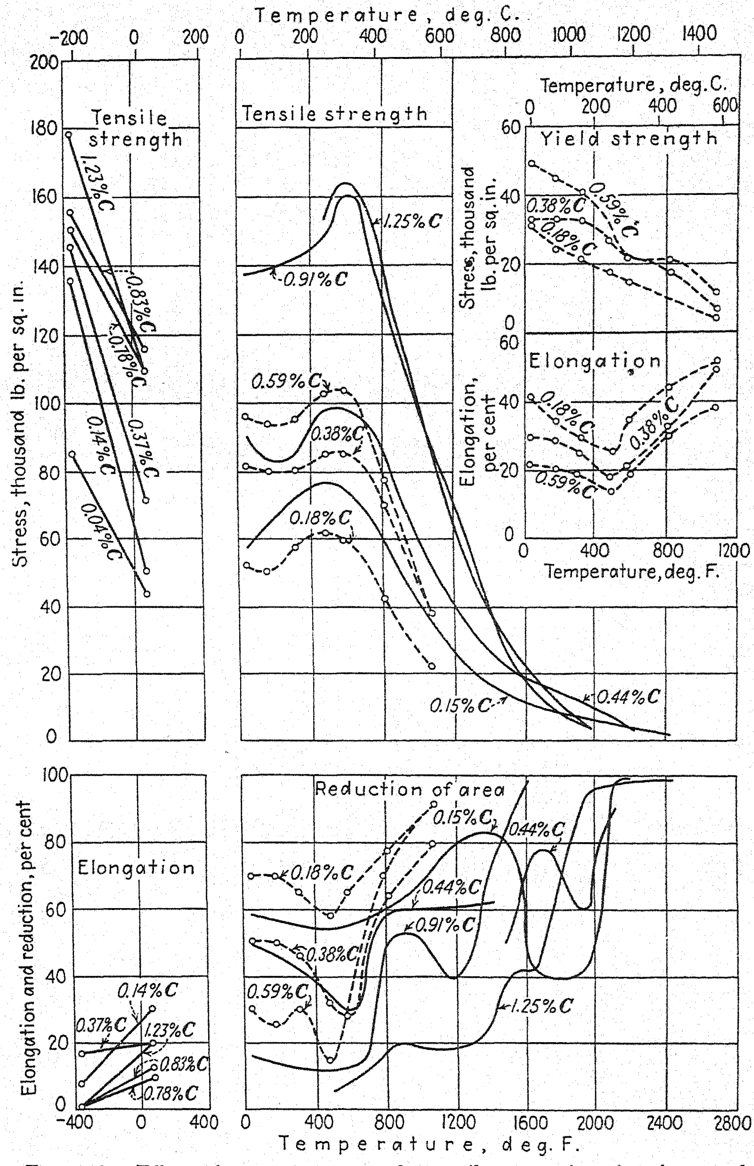

FIG. 140.—Effect of temperature on the tensile properties of carbon steels. Steels used in tests at elevated temperatures were normalized, the ones used at low temperatures were annealed. (*French and Tucker.*[99])

made on slender columns of structural steel from the point of view of fire resistance in buildings. Higher strength was shown at around 250°C. (480°F.) than at room temperature, but above that temperature range the strength decreased rapidly.

The hardness values for several steels and a cast iron at various temperatures, as reported by Brinell,[9] Kürth,[15] and Ito,[100] are shown in Fig. 139. Naturally, the Brinell hardness bears almost a direct relationship to the crushing strength, and the curves shown in Figs. 137 to 139 should show the same trend.

Figure 140, prepared by French and Tucker,[99] shows the tensile properties of several carbon steels at various temperatures. Some of the data shown in the figure were taken from papers by Hadfield[11] and Dupuy.[65]

Values shown in Fig. 139 indicate that it may not be correct to draw smooth curves connecting room-temperature and liquid-air-temperature properties, as is done in Fig. 137. It seems more probable that the curve is flatter just below room temperature and steeper near liquid-air temperature. Figure 137 is in error also in showing the peak in strength at 450°C. (840°F.), for it actually is nearer 300°C. (570°F.), as shown in Fig. 140. That the change in properties somewhat below room temperature is more gradual is also indicated by data from Guillet and Cournot[85] given in Table 83.

TABLE 83.—HARDNESS OF ELECTROLYTIC IRON AND STEELS AT LOW TEMPERATURES*

Material	Brinell hardness at			
	20°C. (70°F.)	−20°C. (−4°F.)	−80°C. (−110°F.)	Liquid-air temperature
Electrolytic iron	80	77	77	269
Soft steel (0.1 % C)	110	107	114	273
Medium steel (0.33 % C)	176	174	190	286
Hard steel (0.79 % C)	230	230	231	330

* Guillet and Cournot.[85]

Johnson and Oberg[624] found that a cold-rolled steel containing 0.28 per cent carbon had the following properties at room temperature and at −40°C. (−40°F.):

Property	Room temperature	−40°C. (−40°F.)
Tensile strength, lb. per sq. in.	89,000	102,000
Yield strength, lb. per sq. in.	68,000	77,000
Brinell hardness	175 to 179	183 to 192

A cold-drawn steel containing 0.46 per cent carbon had the following properties:

Property	Room temperature	−40°C. (−40°F.)
Tensile strength, lb. per sq. in.	175,000	184,000
Yield strength, lb. per sq. in.	129,000	139,000
Brinell hardness	288	306

A general picture of the change in behavior of steels at elevated temperatures is given by Fig. 141 from Welter,[78] which shows complete short-time stress-strain diagrams from room temperature to 500°C. (930°F.), at 100°C. (180°F.) intervals, for four annealed carbon steels of the following analyses:

Element	Percentage			
Carbon	0.37	0.48	0.74	0.85
Manganese	0.76	0.46	0.78	0.25
Silicon	0.4	0.25	0.04	0.3

160. Anomalies at Certain Temperatures.—The data shown in Figs. 137 to 141 were obtained by observations made at temperature intervals too large to bring out clearly the variations which may exist in certain temperature ranges. Abrupt changes occur at the critical ranges, as Sauveur[408] showed. He gave data on short-time tension and torsion tests for steels of different carbon contents in which particular attention had been paid to the higher temperature ranges and to the jogs which occur in the curves at the critical points. Some of the data are shown in Figs. 142 to 146. The torsion tests were made on rods 0.25 in. in diameter; the reduced section had a length of 0.5 in. What is

termed "torsional strength" represents the load required to break the sample and is not the shearing strength. The "factor of stiffness" was obtained by dividing the breaking load by the

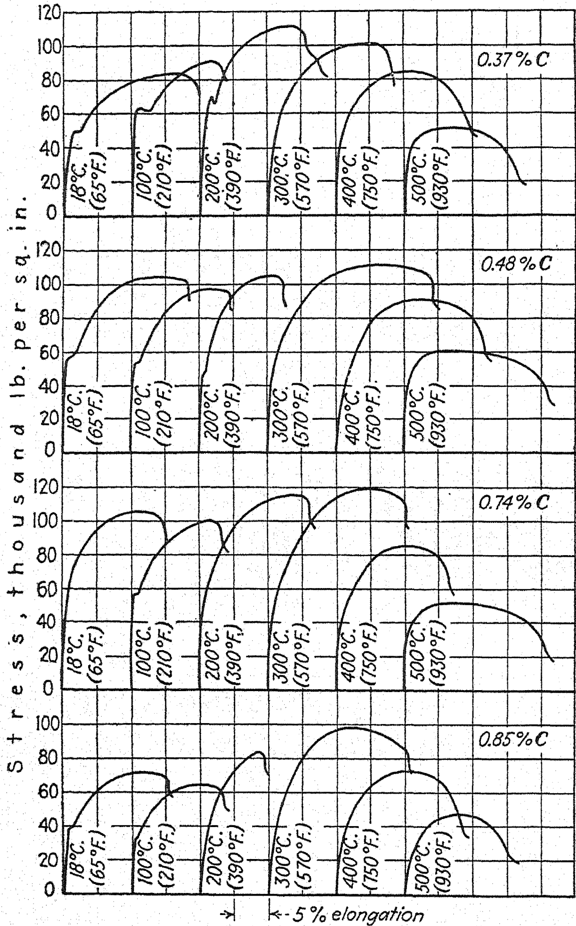

FIG. 141.—Stress-strain curves of four carbon steels tested at elevated temperatures. (*Welter.*(78))

twist, which represents stress in arbitrary units divided by strain. Small irregularities occurring at the critical temperatures have been omitted from Figs. 144, 145, and 146. An earlier paper by Sauveur(137) dealt with the subject in a more qualitative way.

There is a temperature range for material such as ingot iron in which hot working is not feasible. This is just at the lower end of the austenite range. Sauveur pointed out that at a temperature near 900°C. (1650°F.) austenite is stronger and less ductile than is alpha iron just below that temperature, but that with further increase in temperature austenite becomes weaker, more ductile, and is again amenable to hot working. Newell[(810)]

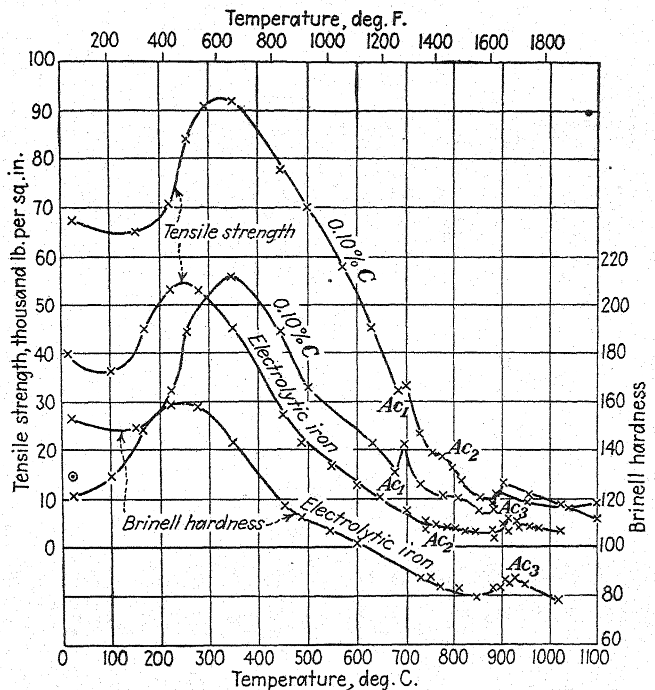

FIG. 142.—Effect of temperature on the tensile strength and Brinell hardness of electrolytic iron and a 0.10 per cent carbon steel. (*Sauveur.*[(408)])

has collected evidence which generalizes this statement since it seems to hold for austenites of widely varying composition. He concluded that improvement in high-temperature ductility may be effected by the addition of elements which tend to form ferrite. He further concluded that, in some of the alloy austenites at least, the higher carbon contents do not impair the ductility of austenite compared with the ductility of comparable austenite which is almost carbon-free. He pointed out that,

while the high-temperature ductility in static tests is improved by so altering the composition that it consists of two phases one of which is ferrite, yet the hot working of the duplex alloy may be attended by difficulties.

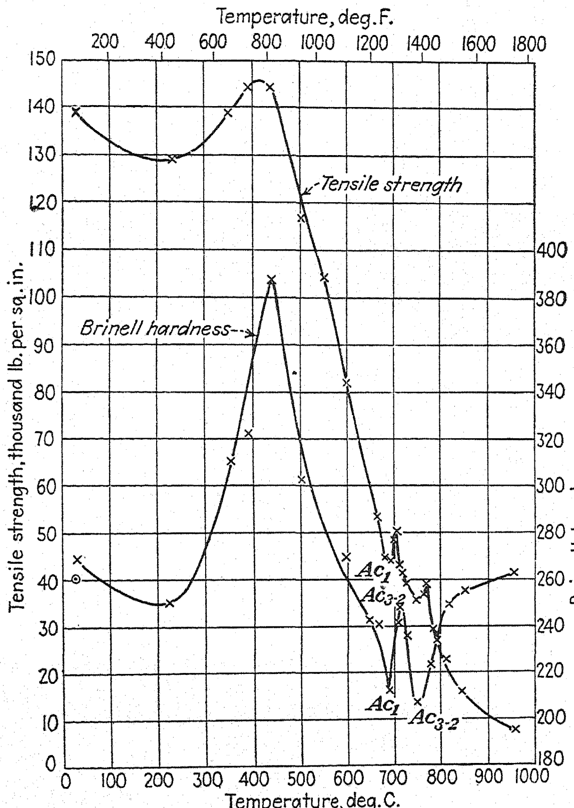

FIG. 143.—Effect of temperature on the tensile strength and Brinell hardness of a 0.75 per cent carbon steel. (*Sauveur.*[408])

The "red-short" or "critical working" range in ingot iron is from 900°C. (1650°F.) to about 1065°C. (1950°F.). With increasing carbon the lower limit of existence of austenite is at lower temperatures. Owing to this and to the effect of carbon in the austenite, hot working is again feasible, so that the red-short range of austenite is met only in very low carbon materials. The presence of certain alloying elements in austenite also tends

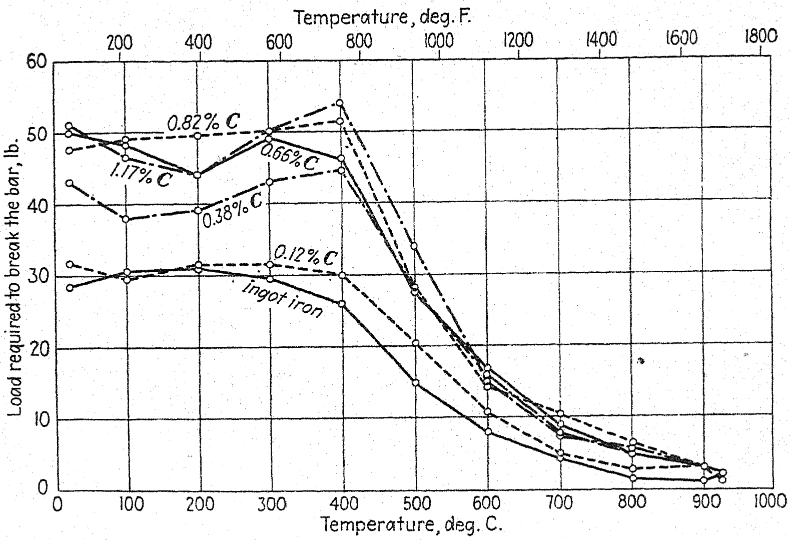

FIG. 144.—Effect of temperature on the torsional strength of ingot iron and carbon steels. (*Sauveur.*[408])

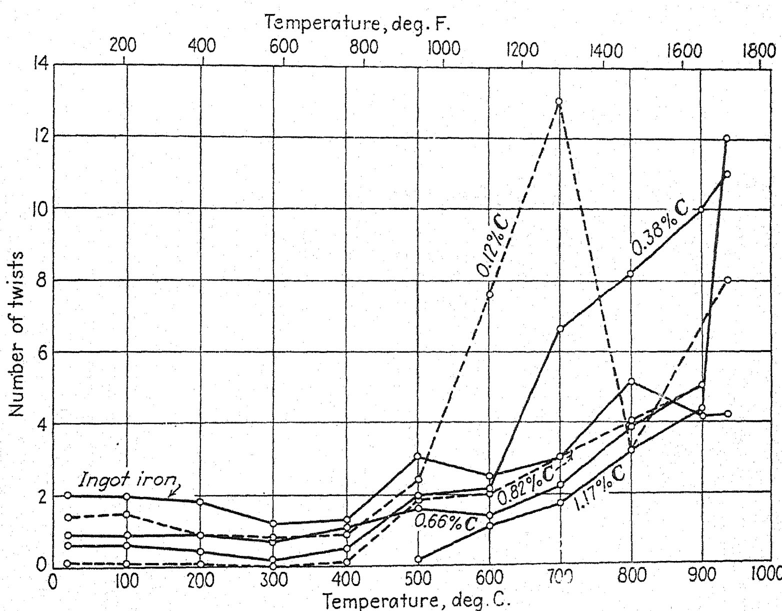

FIG. 145.—Effect of temperature on the number of twists withstood by ingot iron and carbon steels. (*Sauveur.*[408])

to remove hot shortness. Kenyon,[261] however, ascribed the critical working range to the high ratio of sulphur to manganese in ingot iron and stated that with a sulphur content which does not exceed approximately 0.01 per cent the material is workable in this range. Sauveur, in discussion, challenged this explanation, considering the one outlined before as preferable.

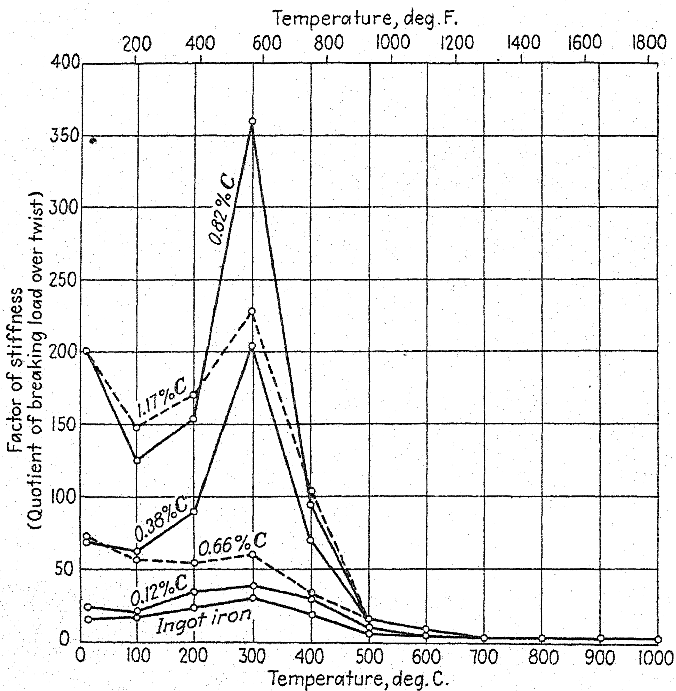

FIG. 146.—Effect of temperature on the factor of stiffness of ingot iron and carbon steels. (*Sauveur.*[408])

161. Elastic Moduli.*—Owing to the large effects of rate of loading and time of application of load at the higher temperatures, neither the location of the proportional limit nor the data for the modulus of elasticity, Young's modulus, are highly exact. Figure 147 from Tapsell[496] shows how the modulus decreases with increasing temperature and how widely the values scatter. The line for "alloy steels" was given by Orrok.[228]

* The modulus of elasticity of iron-carbon alloys at room temperature is discussed in Chapter XV.

Values for 0.40 per cent carbon steel given by Wilhelm[142] agree fairly well with the curve given by Tapsell. All of the carbon

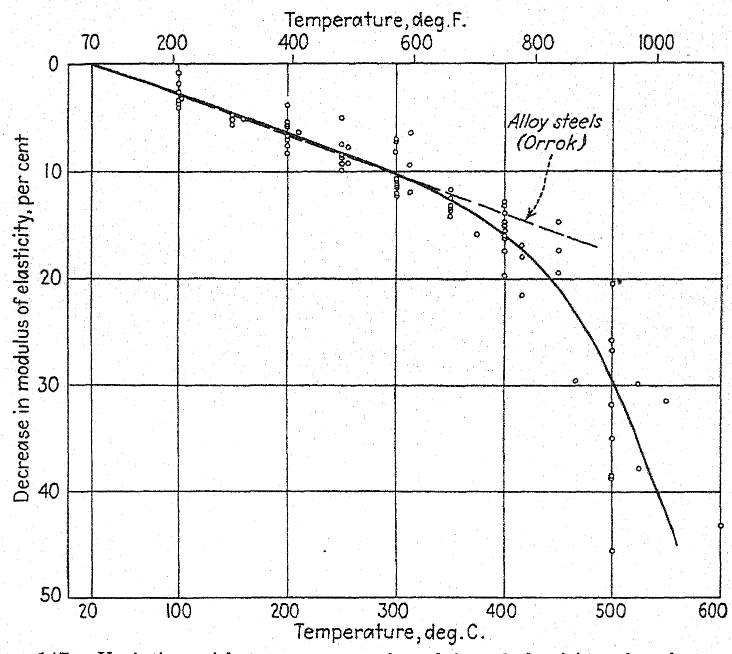

FIG. 147.—Variation with temperature of modulus of elasticity of carbon and low-alloy steels. (*Tapsell.*[496])

steels were averaged by Tapsell and their actual moduli given as follows:

Temperature		Modulus, million lb. per sq. in.
°C.	°F.	
20	70	30.0
100	210	29.2
150	300	28.6
200	390	28.1
250	480	27.5
300	570	26.9
350	660	26.2
400	750	25.2
450	840	23.7
500	930	21.1

The moduli of elasticity of three annealed carbon steels at different temperatures as reported by Welter[78] are given in Table 84. The uncertainty of the values for the higher temperatures is so great that it is not possible to determine from Welter's values whether the rate of decrease in modulus with temperature is influenced by the carbon content.

TABLE 84.—MODULUS OF ELASTICITY FOR SEVERAL CARBON STEELS AT DIFFERENT TEMPERATURES*

Steel number	Composition, per cent			Modulus of elasticity, million lb. per sq. in., at					
	C	Mn	Si	20°C. (70°F.)	100°C. (210°F.)	200°C. (390°F.)	300°C. (570°F.)	400°C. (750°F.)	500°C. (930°F.)
1	0.37	0.76	0.4	28.8	28.2	28.0	26.9	24.3	
2	0.49	1.05	0.58	30.0	32.2	28.5	27.3	21.5	19.9
3	0.74	0.78	0.04	31.2	27.9	29.4	24.9	15.5	12.7

* Welter.[78]

Elastic moduli for carbon steels and ingot iron at elevated temperatures were also given by Lea and Crowther[29] and by Lea.[90] The curve for ingot iron from room temperature to 300°C. (570°F.) showed a smooth decrease in elastic modulus from 29 million lb. per sq. in. to about 22.5 million lb. per sq. in. at the highest temperature.

Honegger[533] recently determined Young's modulus for a medium-carbon steel and several alloy steels over the range from room temperature to 600°C. (1110°F.). He used both static and dynamic methods. At temperatures above 300°C. (570°F.) there was an appreciable difference between values determined statically. Mitinsky[640] studied the modulus of steel up to 800°C. (1470°F.).

Recently reported values of elastic moduli for an annealed medium-carbon steel at temperatures up to 500°C. (930°F.) are shown in Fig. 148 from Versé.[819] The values determined dynamically were considered to be more nearly correct than those determined statically. Poisson's ratio was calculated from values of the moduli. As may be observed from Fig. 148 and the data listed above, Versé's values of a modulus of elasticity are in good agreement with those selected by Tapsell.[496]

Ludewig[736] studied the torsional modulus of a spring steel containing 0.61 per cent carbon, 1.51 per cent manganese, and 0.14 per cent silicon, which had been oil quenched from 830°C. (1525°F.) and tempered at 295°C. (560°F.). At room temperature, the torsional modulus was about 10.8 million lb. per sq. in.; at 360°C. (680°F.) it had fallen to 10.4 million, and from that temperature it decreased linearly to about 9.5 million at 500°C. (930°F.).

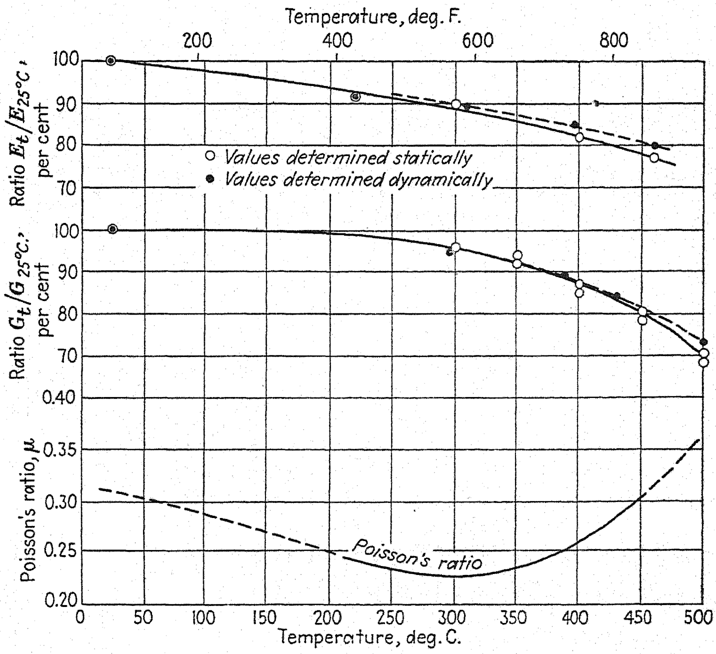

FIG. 148.—Variation of modulus of elasticity E, modulus of rigidity G, and Poisson's ratio μ with temperature, for a medium-carbon steel. (*Versé.*[819])

Temperature coefficients of the modulus of elasticity of several high-carbon steels were recently determined by Keulegan and Houseman.[626] In determining these temperature coefficients they worked in the range -50 to $+50$°C. (-60 to $+120$°F.). The coefficients for the carbon steels were from -20.1×10^{-5} to -27.1×10^{-5}. If the coefficient is -25×10^{-5} and the modulus at 0°C. (30°F.) is 30 million lb. per sq. in., the modulus at -50°C. (-60°F.) would be 30.4 million lb. per sq. in., and at 50°C. (120°F.) it would be 29.6 million lb. per sq. in. Keule-

gan and Houseman also determined the temperature coefficient of the modulus of rigidity for the same steels; it varied from -20.1×10^{-5} to -27.3×10^{-5}. A few data on the effect of subnormal temperatures on the modulus of elasticity are given on page 459.

B. GENERAL TREND OF DYNAMIC PROPERTIES

Most of the values which show the effect of temperature on the impact resistance of carbon steels have been determined at

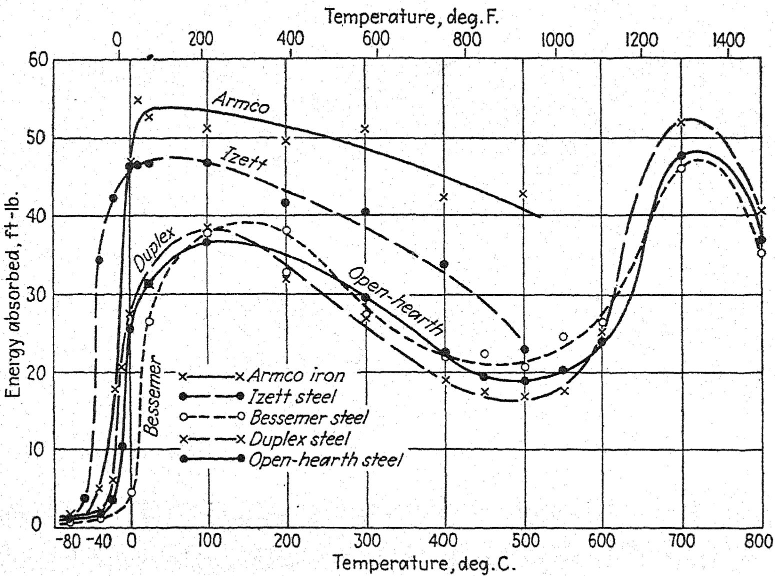

FIG. 149.—Effect of temperature on the Charpy impact resistance of ingot iron and low-carbon structural steels. (*Epstein.*[520])

subnormal temperatures and are, consequently, discussed in sections 165 and 166 (pages 456 and 462). A few data, mostly by Epstein, on high-temperature impact resistance are reproduced in the following section. Few investigations on fatigue properties at temperatures above or below normal have been made, too few to permit any broad generalization of the effect of temperature on this property.

162. Impact Resistance.—Figure 149 from Epstein[520] shows curves of temperature versus impact resistance for several low-carbon structural steels. The steels were in the hot-rolled

condition, and the values shown were obtained by the Charpy test. The precipitous drop in impact values with decreasing temperature in the neighborhood of room temperature is characteristic of many commercial low-carbon steels, as has been shown by other investigators.[126,155]

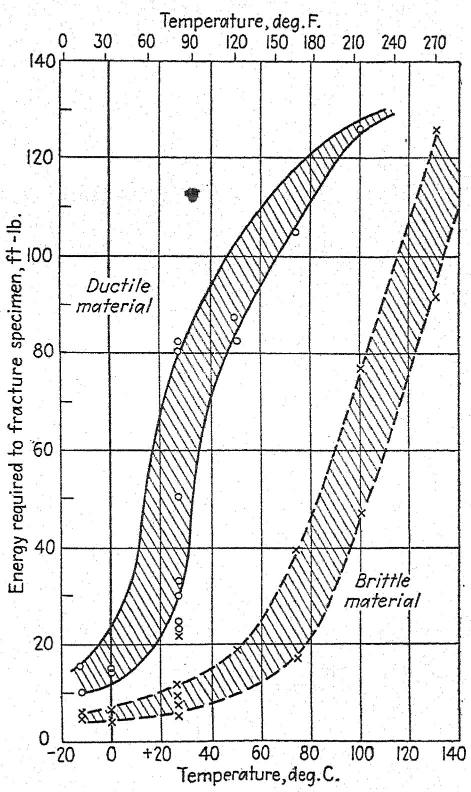

FIG. 150.—Effect of temperature on the Izod impact resistance of brittle (0.08 per cent oxygen) and ductile (0.057 per cent oxygen) ingot irons. (*Hensel and Hengstenberg.*[622])

The fact that materials whose ordinary analysis would not lead one to expect a difference in impact properties may nevertheless show a difference at room temperature is proved by the data of Hensel and Hengstenberg[622] shown in Fig. 150. The curves are for Izod impact values of two specimens of ingot iron of the following analyses:

Specimen	Composition, per cent						
	C	Mn	P	S	Si	O	N
Ductile..........	0.02	0.02	0.005	0.030	0.003	0.057	0.021
Brittle..........	0.02	0.04	0.009	0.043	0.003	0.080	0.004

If an impact resistance of 20 to 40 ft-lb. is taken as the approximate dividing line between brittle and ductile materials of this composition, Fig. 150 shows that the high-oxygen specimens

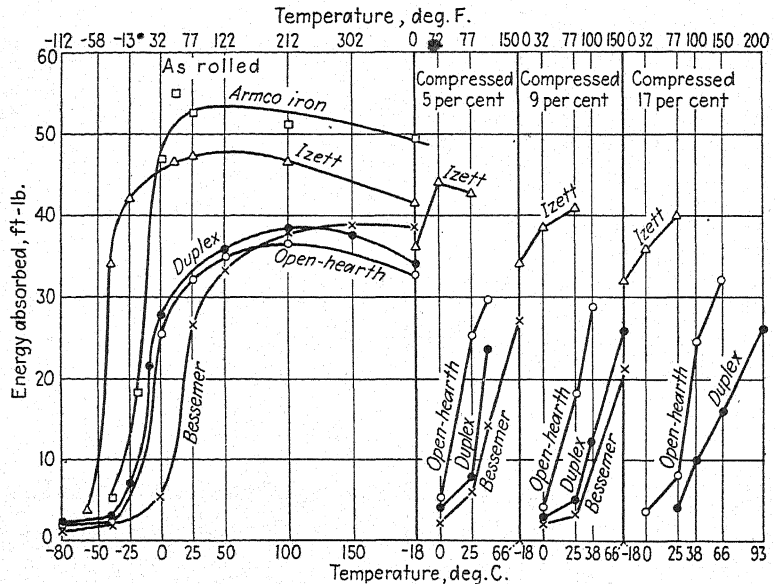

FIG. 151.—Effect of temperature on the Charpy impact resistance of ingot iron and low-carbon structural steels as rolled and as cold compressed. (*Epstein.*[520])

were ductile only at temperatures above 75°C. (165°F.); the lower oxygen specimens, on the contrary, were ductile at all temperatures above normal.

Epstein's[520] data, shown in Fig. 151, indicate that different structural steels show great differences in susceptibility to embrittlement by cold work and that embrittlement is brought about by the shifting of the steep portion of the temperature-impact resistance curves to higher temperatures. Other data reported by Epstein also bring out the great differences in sus-

ceptibility to embrittlement of carbon steels of the structural type. The Izett steel contained 0.10 per cent carbon, 0.55 per cent manganese, 0.04 per cent silicon, 0.05 per cent aluminum, and 0.07 per cent nitrogen. The Bessemer steel contained approximately 0.10 per cent carbon, 0.10 per cent phosphorus, and more nitrogen than the open-hearth steel, while the duplex and open-hearth steels would be indistinguishable on analysis. For the particular heats tested, the duplex and open-hearth

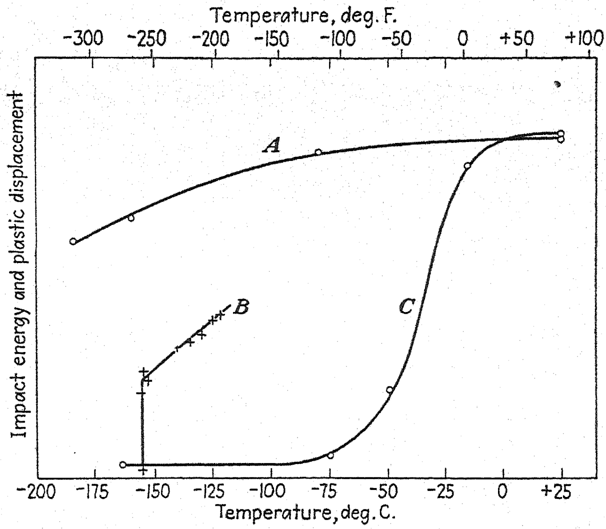

FIG. 152.—Influence of low temperature on (*A*) angular displacement in torsion, (*B*) elongation in tension, and (*C*) energy absorbed in impact; indicating disappearance of plasticity in hydrogen-purified iron at different temperatures depending on the several testing methods. (*Heindlhofer.*[713])

steels differ in impact resistance after cold compression. It is quite possible that the variation in the temperature range at which the single-blow notched-bar impact resistance (notably of cold-worked materials) is greatly lowered is connected with precipitation-hardening phenomena, and that ultimately these variations can be appraised as due to small amounts of adventitious impurities. Heindlhofer[713] stated that there is an abrupt drop in the curve for hydrogen-purified iron with almost complete freedom from impurities, as shown in Fig. 152. It is the shifting about that is thought to be affected. The true effect of the variation in carbon, by itself, on the properties in the impact-embrittle-

ment range and in the blue-heat range (so named because blue temper colors appear on steel at that range of temperatures) cannot yet be evaluated.

163. Endurance Limit.—Because the endurance limit is an important factor in the appraisal of metallic materials where repeated stress is involved, the question arises how the endurance properties change with temperature, although high alternating stresses are not often imposed upon structures or machine parts operating at high temperatures. Little work, however, has been done in this field.

The Joint Research Committee on Effect of Temperature on the Properties of Metals of the American Society for Testing Materials and the American Society of Mechanical Engineers[505] compared in Fig. 153 the results of rotating-beam endurance tests on a 0.17 per cent carbon steel, made by Moore and Alleman of the University of Illinois, with the tensile and creep values for the same steel obtained by Kanter and Spring. The values of endurance limits for temperatures above 425°C. (800°F.) are probably high, as is shown by the arrows. It will be noted that the endurance ratio is as high at 290°C. (550°F.), the temperature of maximum tensile strength, as it is at room temperature. In fact, the endurance ratio is substantially constant for all temperatures investigated. Figure 153 also shows a curve from Wiberg,[580] which indicates a peak in the endurance curve for a 0.15 per cent carbon steel at a higher temperature than that found by Moore and Alleman. High-temperature tensile data for Wiberg's steel are lacking.

Tapsell[496]* gave temperature-property curves for a 0.17 per cent carbon steel. The maximum in the tensile-strength curve occurred at about the same temperature as in Fig. 153, and the maximum in the endurance curve, when the axial stress varied about a mean value of zero, occurred at approximately the same temperature as the maximum found in Wiberg's curve. Under repeated stress varying about a positive (tensile) mean stress of about 22,000 lb. per sq. in., the maximum in the endurance curve shifted to a somewhat lower temperature and, with a mean stress of about 44,000 lb. per sq. in., to a still lower temperature, but not so low as Moore and Alleman's temperature for maximum endurance strength. Tapsell suggested that the

* Page 208 of his book.

displacement of the peak may be a function of speed of loading in fatigue testing, *i.e.*, the rate of reversals. In Tapsell's work at other than zero mean stress the specimens elongated during the test and, at the lower temperatures at least, were work hardened to some degree.

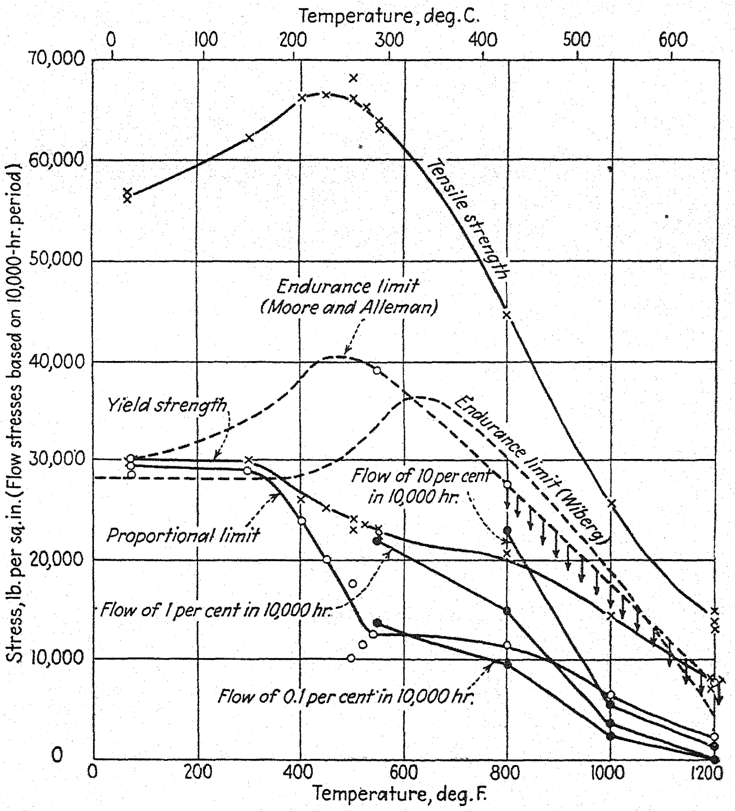

FIG. 153.—High-temperature tensile, creep, and fatigue data for a 0.17 per cent carbon steel, and fatigue data for a 0.15 per cent carbon steel. Curves not marked are from Kanter and Spring. (*Joint Research Committee.*[505])

Jasper[160] showed a curve for a 0.50 per cent carbon steel that indicated very slight net differences in endurance limits from room temperature to 540°C. (1000°F.). There was a slight rise at 425°C. (800°F.), while the maximum tensile strength was at 315°C. (600°F.).

High-temperature endurance values for two annealed carbon steels were given by Kaufmann;[465] his values, however, are somewhat questionable, because tests were not run to more than 10 million cycles even at the highest temperature. While the number is ordinarily sufficient at room temperature, it is not yet proved that it suffices at very high temperatures. The analyses of the steels were:

Number	Composition, per cent				
	C	Si	Mn	P	S
1	0.36	0.33	0.69	0.02	0.03
2	0.62	0.49	0.75	0.02	0.03

The reported properties of these steels were as follows:

Property	Steel number	
	1	2
Tensile strength, lb. per sq. in.	78,000	89,500
Yield strength, lb. per sq. in.	43,000	47,000
Elongation, (l = 10 d), per cent	25.5	22.5
Reduction of area, per cent	60.5	51
Endurance limit, lb. per sq. in., at		
20°C. (70°F.)	40,000	46,000
310°C. (590°F.)	....	51,500
325°C. (615°F.)	53,500	
425°C. (800°F.)	....	40,000
550°C. (1020°F.)	21,000 ±	27,000
710°C. (1310°F.)	....	<13,000

Fatigue tests at elevated temperature on a steel containing 0.35 per cent carbon were reported by Jünger.[463] At room temperature the steel had a tensile strength of 88,000 lb. per sq. in., a yield strength of 52,000 lb. per sq. in., an elongation (l = 10 d) of 21.5 per cent, a reduction of area of 55 per cent, and an impact resistance (Mesnager notch) of 7 m-kg. per sq. cm. For the endurance tests a heating coil was placed about the specimen just below the fillet and a thermocouple inserted in a central hole extending through the lower part of the specimen

nearly to the fillet. No calibration for temperature distribution was given, and there is no certainty that the temperature at the breaking section may not have been considerably lower than that measured, especially at the higher temperatures. The endurance limits given were:

Temperature		Endurance limit, lb. per sq. in.
°C.	°F.	
20	70	42,500
200	390	45,500
300	570	48,500
450	840	45,500
500	930	40,000

The data cited from all these investigations are insufficient to decide definitely whether the endurance ratio will be approximately constant at elevated temperatures. They do, however, indicate that low- and medium-carbon steels up to say 400°C. (750°F.) will have at least an equal, and, at some intermediate temperatures, a higher endurance limit than at room temperature when these limits are obtained by rapidly reversed stresses alternating about a zero mean value. When the stress is not completely reversed, the mode of failure would tend to be by distortion owing to creep as the temperature is increased. Very slow reversal of stress would also be expected to allow creep between cycles and accentuate the importance of creep resistance rather than fatigue resistance. Oxidation at temperatures over 300°C. (570°F.) may render the results of endurance tests at such temperatures unreliable unless they are conducted in a non-oxidizing atmosphere.

Few low-temperature endurance data are on record for carbon steels; among these are those cited by Russell[(484)] from unpublished work at the University of Illinois, in which specimens from a rail head gave an endurance limit of 48,000 lb. per sq. in. at 20°C. (70°F.) and 52,500 lb. per sq. in. at −18°C. (0°F.). Forsman[(611)] made endurance tests, also at −18°C., on a 0.41 per cent carbon axle steel that had been normalized at 840°C. (1545°F.). The endurance limit was slightly higher at −18°C. than at room temperature. These data, and Russell's[(484)] and

Johnson and Oberg's[624] data for a number of alloy steels, indicate that at subnormal temperatures the endurance limit rises proportionately with the tensile strength.

164. Alternate Bending.—In a comprehensive investigation on the effect of temperature on static and dynamic properties, Schwinning, Knoch, and Uhlemann[769] determined tensile properties, impact resistance, and resistance to alternate-bending stresses, with special attention to the effect of notches, on five carbon and five alloy steels. The composition and room-temperature properties of the carbon steels are given in Table 85

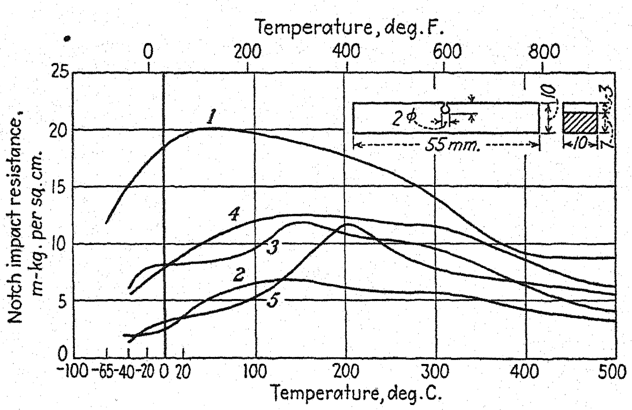

FIG. 154.—Effect of temperature on the impact resistance of carbon steels. The composition and treatment are given in Table 85, page 454. (*Schwinning, Knoch, and Uhlemann.*[769])

Specimens for the short-time tensile tests were heated in an electric salt bath. The yield strength (0.2 limit) and the elastic limit (0.01 per cent permanent set) at elevated temperatures were determined by "the standard procedure of the Verein deutscher Eisenhüttenleute." The resistance to alternate bending was determined by a special machine, in which a rotary bending stress was applied to a smoothly radiused specimen clamped at one end. It was pointed out that this machine yields, on an average, values for alternate-bending strength at 10^7 reversals of stress which are 3 per cent higher than the values obtained by the rotating-beam test.

The results are summarized in Figs. 154, 155, 156, and 157. The first of these, Fig. 154, shows the effect of temperature on the impact resistance, which increases to a peak at a temperature

varying between 50 and 200°C. (120 and 390°F.) depending upon composition and structural condition. A direct comparison between fine- and coarse-grained steels of the same composition is possible; the fine-grained steel has a markedly higher impact resistance at all the temperatures used, although the difference is reduced for temperatures above 400°C. (750°F.); and the peak in the impact resistance occurs at a somewhat higher temperature than for the coarse-grained material. In regard to the impact

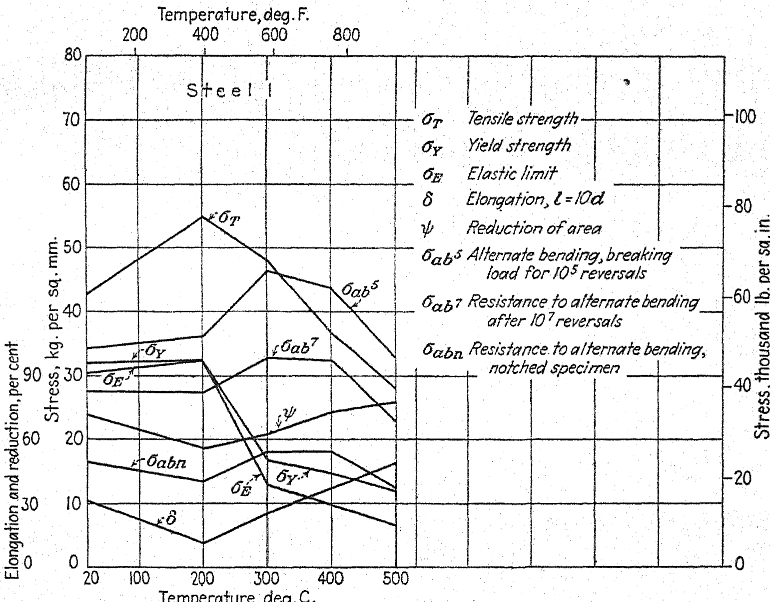

Fig. 155.—Effect of temperature on tensile and alternate-bending strengths of 0.12 per cent carbon steel. See Table 85 for composition and treatment. (*Schwinning, Knoch, and Uhlemann.*[769])

values Schwinning, Knoch, and Uhlemann commented that there is no distinct evidence of blue brittleness in these steels.

From the curves of short-time tensile properties and resistance to alternate bending (Figs. 155, 156, and 157) the following may be concluded:

1. The peak in tensile-strength values occurs at 200°C. (390°F.) for the 0.12 per cent carbon steel—which checks satisfactorily with the Kanter and Spring values given in Fig. 153—and in the other steels at 200 to 300°C. (390 to 570°F.). This

increase in tensile strength is not accompanied by an increase in yield strength. This confirms the results of Kanter and Spring (Fig. 153).

2. The decrease in elastic limit (0.01 limit) with increase in temperature from 300 to 500°C. (570 to 930°F.) is greater than the decrease of yield strength. This is also shown by the Kanter and Spring values for yield strength and proportional limit plotted in Fig. 153.

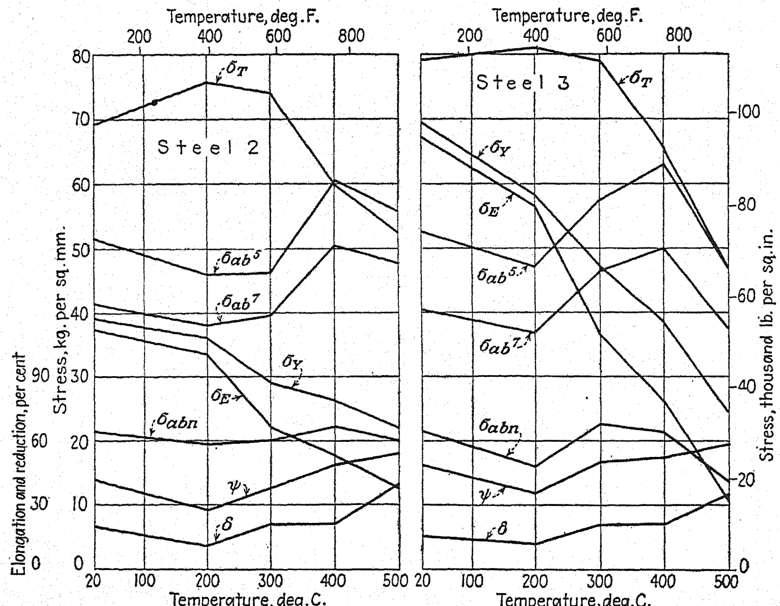

FIG. 156.—Effect of temperature on tensile and alternate-bending strengths of 0.48 per cent carbon coarse-grained (steel 2) and fine-grained (steel 3) steels. See Table 85 for composition and treatment. (*Schwinning, Knoch, and Uhlemann.*(769))

3. The alternate-bending strength for 10^5 and 10^7 reversals increased to a peak at 300 to 400°C. (570 to 750°F.) for the low-carbon steel and at 400°C. (750°F.) for all the other steels except No. 5. The peak for the low-carbon steel (Fig. 155) checks fairly well with the peak in the Wiberg curve in Fig. 153. For all the steels tested by Schwinning, Knoch, and Uhlemann (except No. 5) the peak in the endurance limit was at a temperature 100 to 200°C. (180 to 360°F.) higher than the peak in tensile strength.

4. The alternate-bending strength for all the steels decreased only when testing temperatures above 400°C. (750°F.) were used.

5. The increase in alternate-bending strength at 400°C. (300°C. for steel No. 1) was relatively greater than the increase in tensile strength at 200 to 300°C. As the peak in the endurance curves occurs at a higher temperature than the peak in tensile-

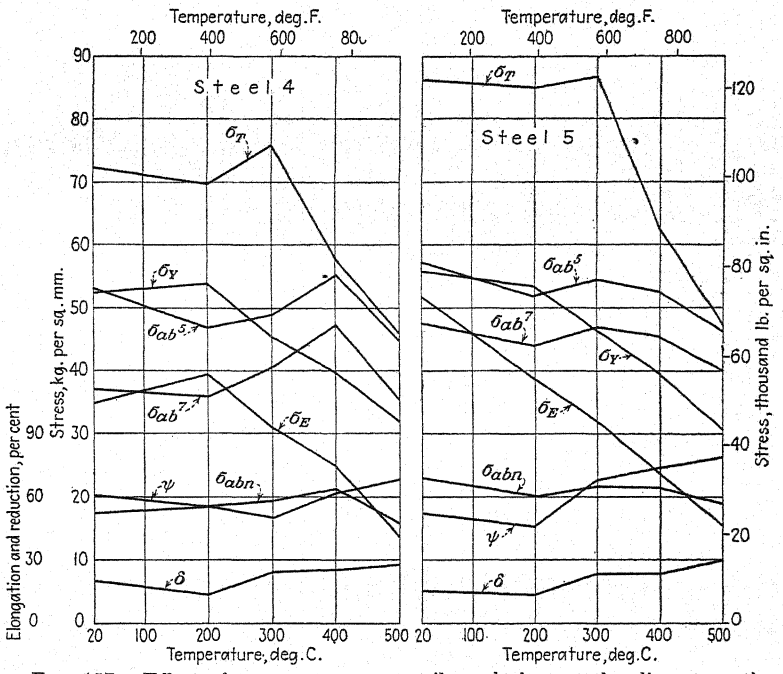

FIG. 157.—Effect of temperature on tensile and alternate-bending strengths of 0.6 per cent carbon (steel 4) and 0.7 per cent carbon (steel 5) steels. See Table 85 for composition and treatment. (*Schwinning, Knoch, and Uhlemann.*[769])

strength curves, it is interesting to note that, in all of the steels except No. 5, at temperatures of 300°C. (570°F.) for steel No. 1 and of 400°C. (750°F.) for the others, the alternate-bending strength for 10^5 reversals is as high as, or higher than, the tensile strength for the same temperature.

6. At 400°C. the value for alternate-bending strength of steel No. 1 is 18 per cent and for steels Nos. 2, 3, and 4, 22 to 27 per cent higher than at room temperature. For all of these steels,

TABLE 85.—COMPOSITION AND ROOM-TEMPERATURE PROPERTIES OF STEELS TESTED BY SCHWINNING, KNOCH, AND UHLEMANN[769]

Steel number	Composition, per cent				Structural condition	Tensile strength, lb. per sq. in.	Yield strength, lb. per sq. in.	Elongation, per cent ($l = 10\ d$)	Reduction of area, per cent	Alternate-bending strength,* lb. per sq. in.	
	C	Mn	Si	Other						Smooth surface	3.5-mm. V-notch
1	0.12	0.52	....	0.18 Cu	As rolled	60,900	41,500	31.1	71	39,300	23,500
2	0.48	0.78	0.35	....	Normalized at 830°C. (1525°F.). Coarse grain	98,400	53,800	19.4	42	58,700	30,300
3	0.48	0.78	0.35	....	As rolled. Very fine grain	112,500	94,000	16.0	49	57,500	30,600
4	0.6	0.6	....	....	Quenched and tempered	102,800	77,400	20.0	61	53,100	24,900
5	0.7	....	....	0.27 Ni	Quenched and tempered	122,300	79,200	15.3	52	67,600	32,700

* The alternate-bending specimen was 241 mm. (9.5 in.) long and 15 mm. (0.59 in.) in diameter. The smoothly radiused section was 95 mm. (3.74 in.) from one end and was 30 mm. (1.18 in.) long with a minimum diameter of 7 mm. (0.28 in.). The notch used was 45 deg., 3.5 mm. deep, with a 0.5-mm. radius.

the value for alternate-bending strength at 300°C. (570°F.) or above is far greater than the yield strength.

7. There are great similarities in elevated-temperature properties in steels Nos. 1 to 4, containing 0.12 to 0.6 per cent carbon. The absence of a well-defined peak in properties, especially alternate bending in steel No. 5 (0.7 per cent carbon), is ascribed by Schwinning, Knoch, and Uhlemann to the presence of 0.27 per cent nickel. There were, likewise, no such peaks in the other alloy steels tested.

8. Between room temperature and 300°C. (570°F.) the tensile strength of the fine-grained steel was somewhat higher than of the coarse-grained material of the same composition (Fig. 156), and the reduction of area somewhat lower. At all temperatures below 400°C. (750°F.) the yield strength and elastic limit of the fine-grained steel were very much higher than the corresponding values for the coarse-grained specimens. There was, however, very little difference between the fine- and coarse-grained specimens in values for alternate bending, and in elongation, for all temperatures below 500°C. (930°F.). At this latter temperature, in tensile strength and resistance to alternate bending the coarse-grained steel was markedly superior to the fine-grained specimen.

9. For all of the carbon steels the effect of a 45-deg. V-notch, 3.5 mm. deep and with a 0.5-mm. radius, was to reduce the resistance to alternate bending about 50 per cent; the values for the notched specimens varied from 44 to 57 per cent of the corresponding unnotched bars. Other forms of notches were used and had relatively greater or less effect depending upon the depth and the radius at the bottom.

C. PROPERTIES AT TEMPERATURES BELOW THE BLUE-HEAT RANGE

In the further correlation of data on the effect of temperature on properties it has been found convenient to divide this discussion according to temperature. Three ranges are recognized. These are: (1) the region below room temperature, (2) the region from room temperature to the blue-heat range, about 300°C. (570°F.), and (3) the region above the blue-heat temperature. In the following pages the first two ranges are considered;

discussion of data on the blue-heat range is postponed until later (page 472).

Room temperature affords a good dividing line between the first two regions because the strength at room temperature is about the lowest found between the temperature of liquid air and approximately 300°C. (570°F.). Fortunately, the change in properties, with the exception of notched-bar impact resistance, is so little for small variations near room temperature that no special attention need be paid to the exact "room" temperature

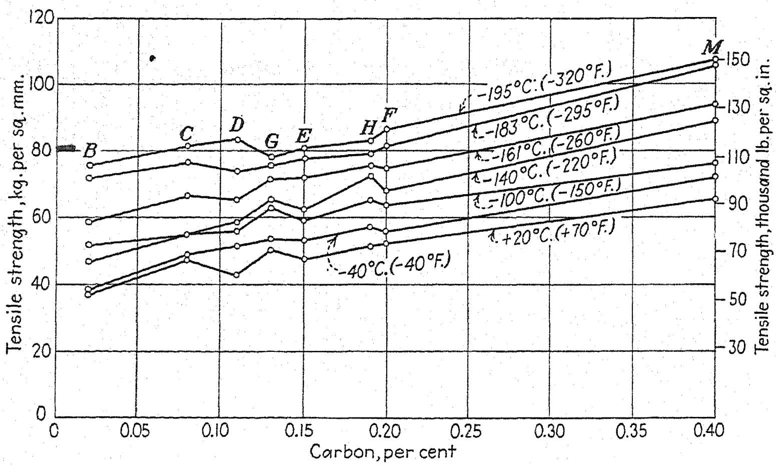

FIG. 158.—Effect of carbon content on low-temperature tensile strength. Carbon percentages are indicated by the plotting points; manganese contents beginning with *B* were 0.05, 0.35, 0.47, 0.60, 0.48, 0.71, 0.52, and 0.80 per cent. (*Gruschka.*[708])

when testing. Nevertheless, a survey of the whole region down to liquid-air temperature shows, as Russell[484] brought out, that tensile and yield strengths, hardness, endurance limit, and modulus of elasticity increase as temperature falls from room temperature, while elongation and reduction of area tend to decrease and notched-bar impact resistance, especially in carbon steels, generally falls sharply. Strauss, in discussion of Russell's paper, pointed out that the changes outlined by Russell may not always occur, as is proved by some of Sykes's[141] data on alloy steels at low temperatures.

165. Properties at Subnormal Temperatures.—The general trend of increasing strength and hardness as the temperature

falls below normal room temperature is shown by the results of many observers. Russell[484] has summarized these and given

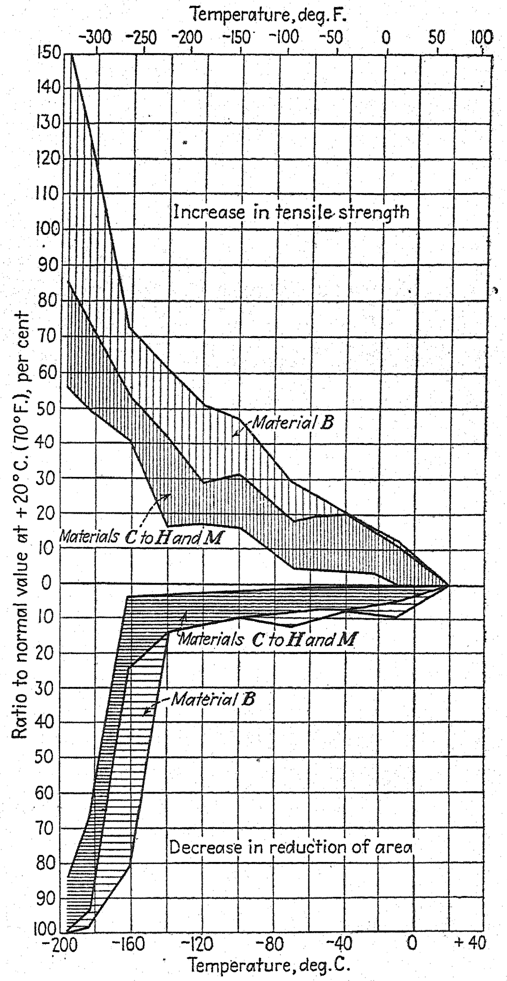

Fig. 159.—Ranges of values for increase in tensile strength and decrease in reduction of area in carbon steels at low temperatures. See Fig. 158 for composition. (*Gruschka.*[708])

references to the earlier literature. The subject has been recently discussed by several writers.

Gruschka[708] studied the ability of steels to withstand static deformation at very low temperatures by tensile tests at tem-

peratures down to −195°C. (−320°F.), obtained by liquid air, oxygen, or nitrogen. He used hot-rolled carbon steels of from 0.02 to 0.40 per cent carbon. Some of his data are shown in Figs. 158 and 159.

The 0.08 and 0.11 per cent carbon steels (materials *C* and *D*) had more ductility remaining at −195°C. (−320°F.) than the higher carbon steels, but the 0.15 per cent carbon steel (material *E*) was no better than those of still higher carbon. Ingot iron showed a greater percentage increase in strength at the very low temperatures than did the steels, the drop in ductility came at a higher temperature, and at −195°C. (−320°F.) it was statically the least ductile of any of the materials tested.

No impact determinations were made in this investigation. Dynamic toughness in the carbon steels would be lost at higher temperatures than is shown by Gruschka's static data. This can be seen from Fig. 152 taken from Heindlhofer.[713]

Thirty years ago Hadfield[11] made a comprehensive study of the properties of steels at liquid-air temperature, and De Haas and Hadfield[600] have supplemented it by a recent investigation.

While the strength increases as the temperature falls, the ductility decreases and the impact resistance falls markedly. The differences in impact resistance among steels of quite similar composition (shown in Fig. 151) at only moderately subnormal temperatures, such as those encountered in winters in northern latitudes or in aircraft flights at high altitude, make the selection of steels for service at those temperatures an important matter.

An investigation of the properties of a medium- and a high-carbon steel was recently completed by the National Bureau of Standards.* Tests were made at room temperature, −78.5°C. (−110°F.), and at some intermediate temperatures on steels of the following compositions:

S.A.E. number	Composition, per cent				
	C	Mn	Si	S	P
1045	0.45	0.77	0.21	0.022	0.013
1095	0.93	0.27	0.15	0.023	0.014

* Louis Jordan, private communication, Feb. 27, 1935. Data published by permission of the Bureau of Aeronautics.

The material was received in the form of cold-drawn and of mill-annealed bars, ⅝ in. in diameter. The mill-annealed bars were subjected to the following treatments:

Steel 1045. Normalized at 870°C. (1600°F.) and air cooled. Quenched in water from 800°C. (1475°F.) and tempered at 540°C. (1000°F.). Air cooled after tempering.

Steel 1095. Normalized at 800°C. (1475°F.) and air cooled. Quenched in oil from 775°C. (1425°F.) and tempered at 315°C. (600°F.). Air cooled after tempering.

Tensile (0.2525 × 2-in. specimens), Rockwell *C* hardness, and Charpy impact (10 mm. square by 55 mm. long, V-notch 2 mm. deep) tests were made at room temperature and at the subnormal temperatures shown by the plotting points in Figs. 160 and 161. Yield strength was determined by drop-of-beam for the 1045 steel and from the 0.2 limit for the 1095 steel. The crosses in Figs. 160 and 161 are values for specimens which had been cooled to −78.5°C. (−110°F.), allowed to return again to room temperature, and then tested at room temperature.

The results for the medium-carbon steel (Fig. 160) confirm other data discussed above, that tensile strength, yield strength, and hardness are higher at approximately −80°C. (−110°F.) than at room temperature, but, as would be expected from the results of Gruschka (Fig. 159), the increase is not great. Reduction of area, as also shown by Gruschka, is somewhat lower at −80°C. than at room temperature. Elongation values are but slightly affected by the low temperature; in the cold-rolled and quenched and tempered specimens they are a little above the room-temperature values, in the normalized specimen a little below. Charpy impact specimens are, of course, deleteriously affected by low temperature; the effect is greatest in the quenched and tempered specimen with a high room-temperature impact resistance. The tensile properties and hardness of the high-carbon steel follow the same general trends (Fig. 161) as in the medium-carbon steel, but the Charpy impact value, already very low at room temperature, is erratic and is apparently but little changed at −78.5°C.

The effect of cooling to −78.5°C. (−110°F.) on the modulus of elasticity of these two steels is noteworthy. In the medium-carbon material it is appreciably lower and in the high-carbon specimens unchanged, as compared with the modulus at room temperature.

While it is probable, as shown later, that steels of small grain size will in general tend to retain their good impact properties to fairly low temperatures, it has been considered scarcely possible to insure a high degree of toughness at extreme subnormal

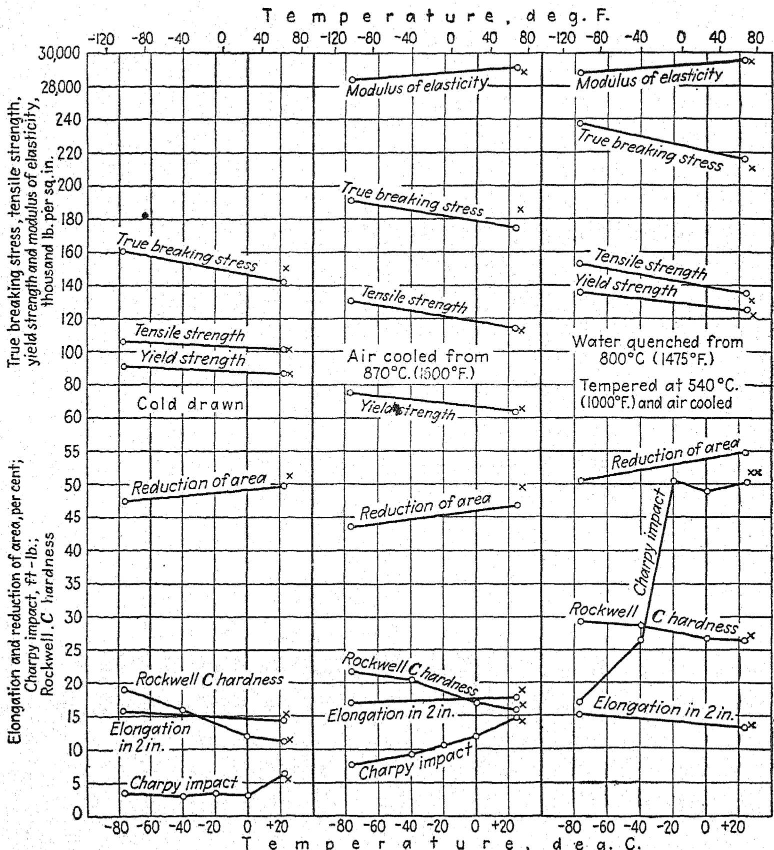

FIG. 160.—Low-temperature properties of 0.45 per cent carbon, 0.77 per cent manganese, 0.21 per cent silicon (1045) steel. Crosses are values for specimens cooled to −78.5°C. (−110°F.), allowed to return to room temperature, and tested at room temperature. (*National Bureau of Standards.*)

temperatures in carbon steels as a class and, in general, alloying must be resorted to to secure such toughness.

At moderately low temperatures, such for example as those used in some processes for dewaxing petroleum products, and the very low temperatures used in making liquid air, carbon steels

as a class become very definitely brittle, and for equipment which must withstand shock alloy steels are chosen. Nickel steels are outstanding in their low-temperature toughness, and a variety of

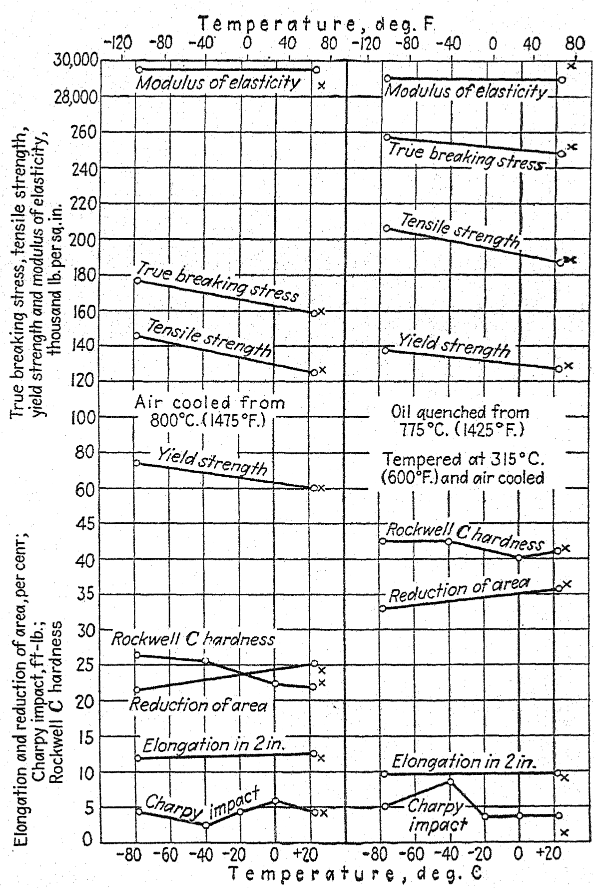

FIG. 161.—Same as Fig. 160 for steel 1095 containing 0.93 per cent carbon, 0.27 per cent manganese, and 0.15 per cent silicon. (*National Bureau of Standards.*)

low-alloy steels is being developed which show promise of toughness at these temperatures. A number of important investigations on low-temperature properties of carbon and alloy steels have been published.*

* See bibliography references 77, 85, 101, 102, 109, 126, 141, 155, 345, 371, 394, 409, 488, 515, 563, 570, 577, 607, 631, 662, 712, 720, 768.

166. Effect of Grain Size and Impurities on Low-temperature Properties of Carbon Steels.—The results of investigations by Herty and McBride[717] and Herty[801] on the effect of deoxidation on impact resistance of carbon steels at low temperature are striking and demand detailed attention. In these investigations, the effect of heat treatment, in addition to the effect of deoxidation, on grain size and thus on the resulting low-temperature impact properties was studied.

The impact specimen chosen had the V or Izod notch (0.01-in. radius) in a bar otherwise of the Charpy type (0.394 in. square and 2.165 in. long). The V-notch was chosen as more likely to discriminate between degrees of toughness, giving lower values than the round-notch Charpy bar on brittle materials and higher values than the standard Izod bar on tough materials.

Six low-carbon and two medium-carbon heats were tested. Their compositions are given in Table 86. It will be noted that

TABLE 86.—ANALYSIS OF STEELS USED BY HERTY AND McBRIDE[717]

Heat number	Composition, per cent								
	C	Mn	P	S	Si	Al*	Ni	N_2	O†
SMA-15	0.15	0.71	0.010	0.029	0.074	0.025	....	0.005	0.0012
SMA-17	0.17	0.56	0.026	0.028	0.150	0.038	0.26	0.003	0.0023
SMA-22	0.22	0.58	0.020	0.043	0.088	0.033	....	0.004	0.0015
K-7	0.15	0.56	0.024	0.040	0.074	0.010	....	0.005	
AX	0.21	0.43	0.016	0.043	0.111	.....	....	0.005	
R	0.14	0.51	0.021	0.035	.....	.....	....	0.007	
SMA-48	0.48	0.74	0.022	0.030	0.205	0.020			
AL	0.40	0.65	0.031	0.029	0.175				

* Metallic.
† Total oxygen, determined by the vacuum-fusion method.

one steel contained 0.26 per cent nickel, an element known to have, at least in larger amounts, a beneficial effect upon low-temperature impact. The six low-carbon heats included one rimmed steel (heat *R*), one silicon-killed steel (heat *AX*), one semikilled steel (heat *K*-7), and three killed steels (heats *SMA*-15, 17, 22). The medium-carbon heat *SMA*-48 was made by the same method as the low-carbon *SMA* steels. Heat *AL* was killed by high-silicon spiegel.

The steels designated *SMA* were killed with silicomanganese and aluminum, and residual metallic aluminum, above that required to combine with the oxygen present, is shown by the analyses. They were fine grained. At sufficiently high tem-

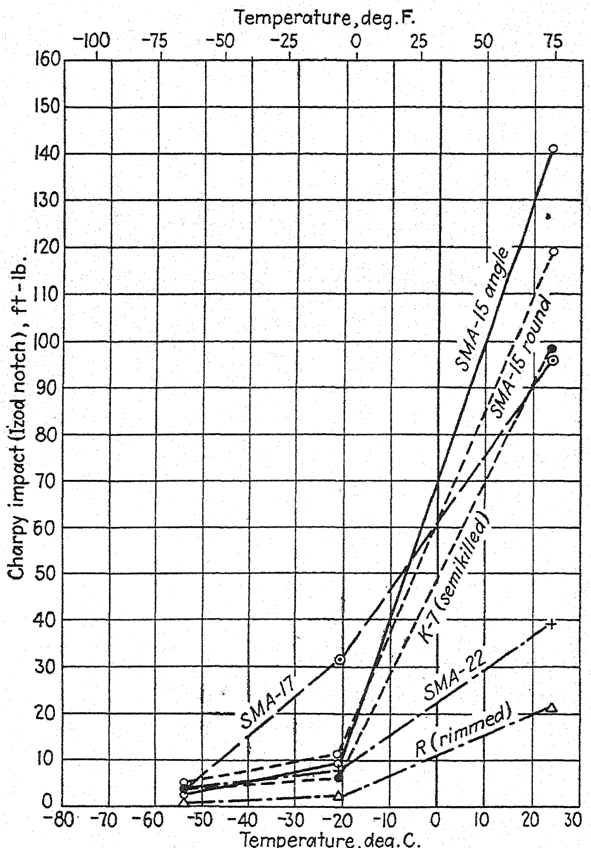

FIG. 162.—Effect of deoxidation on low-temperature impact resistance of rolled low-carbon steels whose composition is given in Table 86. Specimens for heat *SMA*-15 were taken from 1-in. angles and 1-in. round bars. Specimens for other heats were from 1-in. round bars. (*Herty and McBride.*[717])

peratures grain growth can, of course, be obtained even in such steels, so the program included treating these steels to produce coarsening. The rimmed steel (heat *R*) showed somewhat erratic behavior on normalizing, and the results obtained on it may not be characteristic of all rimmed steels.

The impact properties at room temperature, −19°C. (−2°F.), and −54°C. (−65°F.) of the low-carbon steels as rolled are shown in Fig. 162. The beneficial effect of the nickel is apparent. With the exception of the nickel-bearing steel, all lost toughness

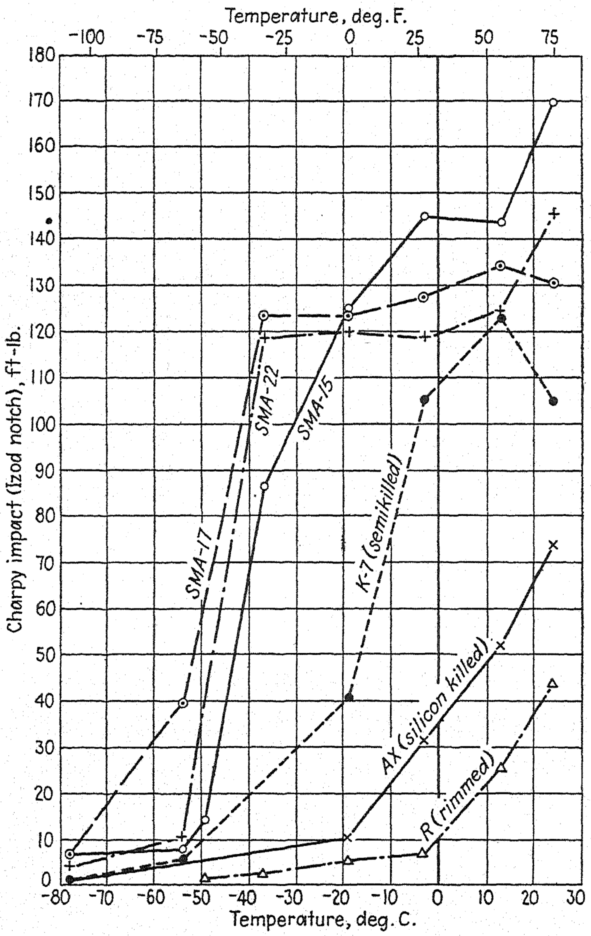

FIG. 163.—Same as Fig. 162, except that the steels were normalized at 875°C. (1605°F.). (*Herty and McBride.*[717])

rather abruptly around −18°C. (0°F.), but the finer grained steels were not so brittle as the coarse-grained ones.

Normalizing—from 875°C. (1605°F.)—produced a very different picture, as Fig. 163 shows. Toughness of the fine-grained

steels was retained at −37°C. (−35°F.) and the nickel-bearing steel retained it to still lower temperatures. The unkilled, silicon-killed, or only semikilled steels fell into another and inferior class. Increasing the normalizing temperature to 1000, 1050, and finally to 1100°C. (1830, 1920, and 2010°F.), to produce coarsening, shifted the curves. Some detrimental

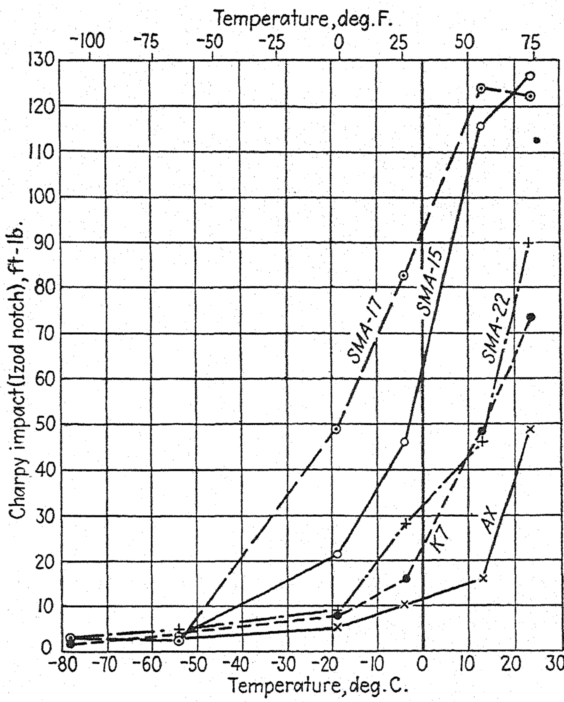

FIG. 164.—Same as Fig. 162, except that the steels were heated to 1100°C. (2010°F.) and air cooled. (*Herty and McBride.*[717])

effect of overheating was noted even before coarsening began—which was taken to indicate that increased solubility of nonmetallics is being brought about and exerts an influence. The impact values after treatment at the highest normalizing temperature are shown in Fig. 164. Smaller differences were shown in the normalized aluminum-killed and silicon-killed medium-carbon steels, which were harder and less tough at ordinary temperatures, as is shown in Fig. 165.

Quenching and tempering greatly benefited the steels, with the exception of the silicon-killed steel (Fig. 166). Many further details are given by Herty and McBride. Herty[801] also showed that the aging characteristics of the steels are wholly

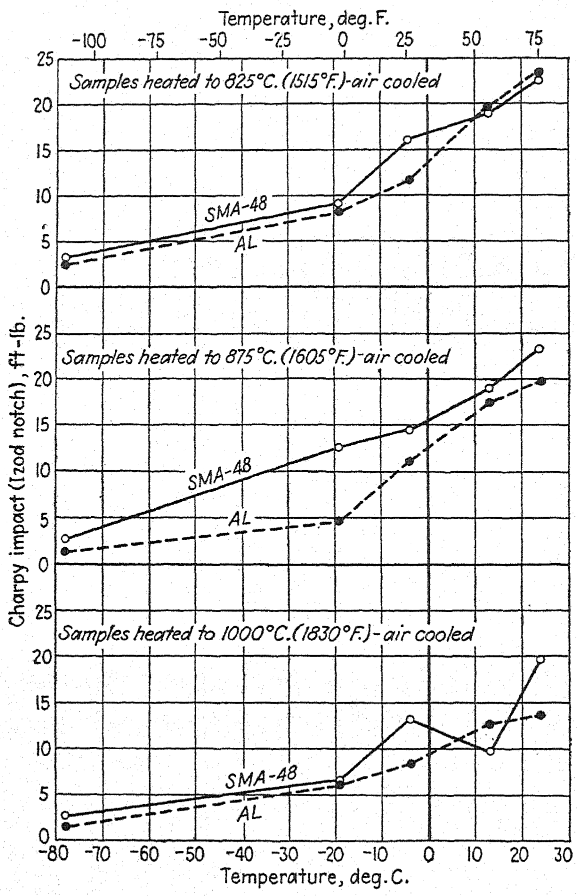

FIG. 165.—Effect of deoxidation on the low-temperature impact resistance of medium-carbon steels. See Table 86 for composition. (*Herty and McBride.*[717])

consistent with their low-temperature impact properties. Resistance to aging goes hand in hand with good low-temperature behavior.

These results show clearly that grain size is of vast importance in its bearing on low-temperature impact properties, exerting so strong an influence that the alleged specific effects of small

amounts of alloying elements, as determined by various investigators, might readily be outweighed by variations in the grain size which are not due to the alloy additions, but merely to the method of deoxidizing the steel.

Before the true effect of manganese, silicon, chromium, copper, etc., can be appraised, it will be necessary to make alloying

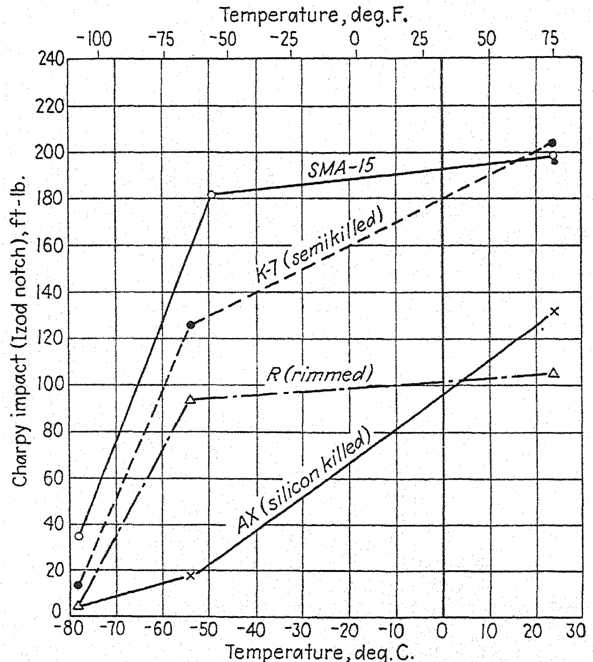

FIG. 166.—Effect of deoxidation on the low-temperature impact resistance of low-carbon steels, water quenched from 875°C. (1605°F.) and tempered at 675°C. (1250°F.). See Table 86 for composition. (*Herty and McBride.*[717])

additions of these elements to carbon steels so deoxidized that they have different grain sizes. It is quite possible that, as in the case of nickel which clearly exerts a specific effect, other elements may have helpful effects when alloyed with a steel which has been made fine grained, as by the use of aluminum or vanadium; but until this has been shown, no reliable conclusions can be drawn.

The work of Herty and McBride has given a new starting point from which it should now be possible to proceed with correct evaluations of the effects of alloying elements.

Schmidt[766] studied the effect of impurities in iron on low-temperature impact properties. He refined Armco iron, by a

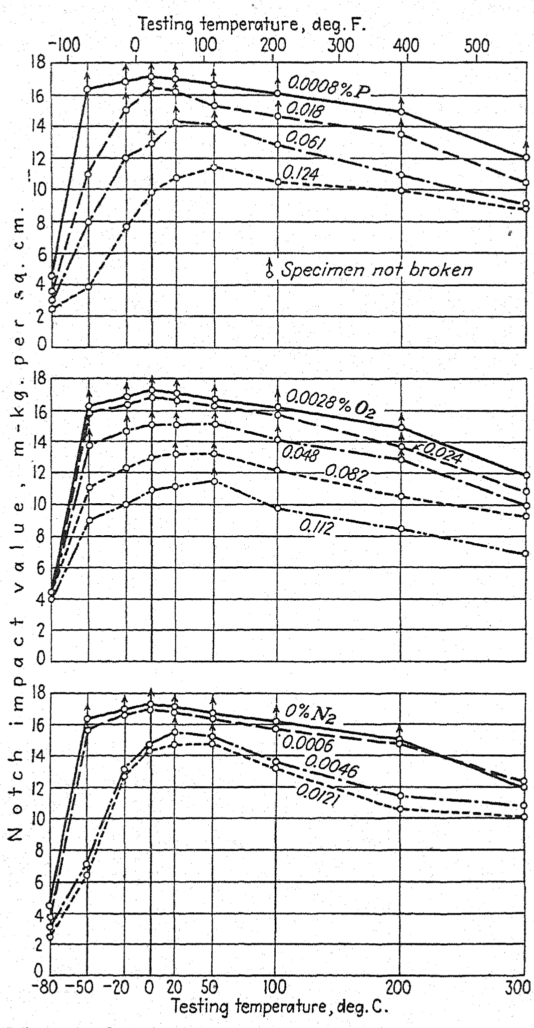

FIG. 167.—Effect of phosphorus, oxygen, and nitrogen on the high- and low-temperature impact resistance of high-purity ingot iron. (*Schmidt.*[766])

series of slagging reactions, till his analyses indicated that 0.001 per cent carbon, 0.008 per cent manganese, 0.002 per cent sulphur, 0.028 per cent oxygen, and no nitrogen were present.

He then added in separate series carbon, phosphorus, sulphur, oxygen, sulphur plus oxygen, and nitrogen (the last by melting under nitrogen pressure). Some of his curves for annealed specimens are shown in Fig. 167. His curves (not here shown) for carbon indicated that this element had about the same effect as does phosphorus, while sulphur alone, up to 0.08 per cent, or oxygen plus sulphur, up to 0.05 per cent oxygen and 0.06 per cent sulphur, had practically no effect. As Fig. 167 shows, neither ordinary amounts of phosphorus nor even large amounts of oxygen brought about an abrupt lowering of the impact values at temperatures as low as −50°C. (−60°F.), but rather small amounts of nitrogen caused a sharp drop between −20 and −50°C. (−5 and −60°F.).

The effect of these impurities on embrittlement, after cold work and artificial aging, was different from their effect on annealed specimens. Neither phosphorus nor nitrogen appeared to have much effect, but oxygen, sulphur, or oxygen and sulphur together had an increasingly detrimental effect as their amounts increased.

167. Properties in the Temperature Range from Normal to 300°C. (570°F.).—As discussed in a following section (page 472), when the temperature is raised from normal to about 300°C. (570°F.)—the blue-heat temperature—the strength, hardness, and endurance limit rise to a peak, while elongation, reduction of area, and notched-bar impact resistance fall. The peak in strength and hardness at the blue-heat temperature may not be a true phenomenon of the iron-carbon system at all, for as, among others, Neuendorff[(554)] and Mehl and Briggs in discussion[(498)] have shown, low-carbon Izett steel does not show this peak, and it is, at the same time, practically free from both aging and the blue-heat phenomena. Apparently, the aluminum is introduced at such a stage in the process of deoxidiation that it combines with and renders insoluble in the ferrite all of those elements whose presence causes the phenomena of aging and blue brittleness. As the aluminum is present in very small quantities and is without doubt largely in the combined state, as inclusions, Izett steel corresponds to the definition of carbon steel used in this monograph. Galibourg[(613)] attributed the peak in the tensile-strength curve at the blue-heat temperature to aging action in the strained steel during tensile testing.

After the blue-heat temperature, 300°C. (570°F.), has been passed, a range of steadily decreasing tensile and yield strengths, hardness, endurance limit, and modulus of elasticity is entered, and plastic behavior becomes very evident in the higher portion of this range. In general, there is a marked trend toward increase in ductility and notched-bar impact resistance as the temperature rises, but this general trend is subject to important exceptions in the case of individual steels, for at 450 to 600°C. (840 to 1110°F.) some steels may show the phenomenon of secondary brittleness. The trend of these properties in the temperature range between normal and 300°C. (570°F.) is shown in Figs. 140, 142, 149, 153, 154, 155, 156, 157, and in Figs. 168 and 174, which are discussed in the next section.

168. Effect of Factors Other than Composition on Properties as a Function of Temperature.—Most of the discussion given so far in the present chapter has been concerned with the relation between properties and temperature for steels of different carbon content. Enough data have been given, however, (for example on effect of grain size and impurities) to show that there are individual idiosyncrasies not ascribable to carbon content which may crop up in carbon steels, especially in relation to impact properties, to disturb the general trend of curves showing properties as a function of temperature. Differences in degree of oxidation, methods of finishing the heat of steel, the presence of adventitious impurities that allow precipitation hardening, or the re-solution of precipitates at elevated temperatures so complicate the behavior with respect to impact resistance, zones of higher strength or lower ductility, short-time tensile and probably creep properties that each particular lot of steel must be to some degree a law unto itself in regard to its high-temperature behavior. Better controllable factors whose effects must be considered are grain size and metallographic structure, which may be at least partially controlled by heat treatment.

The property-temperature curves of annealed, stable steels of different carbon contents have much the same shape but start from different levels corresponding to the carbon content. However, above the recrystallization temperature these differences tend to be minimized until just below the eutectoid temperature, where carbon content ceases to be such a potent factor, though still exerting a detectable influence; in the austenitic

range the differences in strength due to carbon content are small in absolute magnitude, though sufficiently great to be detectable in hot working.

Through the blue-heat range and up to somewhat past the recrystallization temperature the strength and structure produced by quenching and tempering may, within limits, be preserved. Tempering is fundamentally a time-temperature phenomenon, and a steel which has been tempered for say 1 hr. at 260°C. (500°F.) may soften further in 1000 hr. at that temperature. The properties shown by a 1000-hr. test will then not be the true stable properties at 260°C. (500°F.), but the resultant of initial unstable properties and of the more stable ones. If the temperature is then raised to 425°C. (800°F.), the steel previously heated for 1000 hr. at 260°C. (500°F.) will be further softened. Hence, the only permanent high-temperature properties are those of steels made truly stable for the temperature in question by annealing, normalizing and tempering, or by quenching and tempering, with the tempering temperature high enough and the time long enough to insure no change in structure at the lower, though elevated, temperature of test or service.

Of course, the effect of cold work will similarly be removed by heating to the softening or recrystallization temperature, and short-time tests will obviously not indicate whether the material is in a truly stable state; stability must be established by other criteria. The benefits of quenching and tempering shown by room-temperature tests may appear to be retained in whole or in part on a short-time test, but they may not exist in actual long-time service at high temperatures. The stability of properties of quenched and tempered steels is of greatest importance at relatively low temperatures, such as temperatures to which bolts are subjected in steam-plant equipment, where the added strength and resistance to tempering of alloy steels make them the usual choice over carbon steel.

Spheroidization of cementite at temperatures between 400°C. (750°F.) and the lower critical temperature may occur with a resulting decrease in long-time load-carrying ability. Annealing prior to exposure to temperature in this range, according to Tapsell,[(496)]* delays spheroidization and raises the temperature

* Page 192 of his book.

at which it will occur in a given time. A coarse grain structure in cast steel also helps to retard spheroidization. Bailey and Roberts,[507] as discussed in the next chapter (page 520), showed that in the stable spheroidized condition a carbon steel had a greater creep rate.

Normalized, normalized and tempered, or annealed carbon steels are now considered to be in more reliable shape to resist long-time loading at high temperatures than those in a less stable condition.

D. THE BLUE-HEAT RANGE AND THE RANGE OF SECONDARY BRITTLENESS

In the blue-heat range, near 300°C. (570°F.), most steels are stronger and less ductile than at slightly higher or slightly lower temperatures. In the range of secondary brittleness, near 500°C. (930°F.), those steels that exhibit secondary brittleness are less ductile than at somewhat higher or somewhat lower temperatures. Neither range corresponds to any known transformation in the iron-carbon system, and it is at least possible that *pure* iron-carbon alloys would not exhibit either blue brittleness or secondary brittleness.

169. The Blue-heat Range.—Figure 168 from Inokuty[259] shows how the tensile strength of Armco iron and annealed carbon steels varies as the temperature is raised to 700°C. (1290°F.). The data on notched-bar impact resistance of several carbon steels at elevated temperatures reported by Reinhold[41] and Guillet and Révillon,[14] plotted in Fig. 169, show that a minimum impact resistance occurs near the temperature at which a maximum tensile strength is observed—around 300°C. (570°F.). Figure 170 from a Committee *Report*,[583] shows the short-time tensile properties of a 0.28 per cent carbon steel in both the cast and wrought conditions.

Since the ductility drops as the tensile strength rises, the strengthening at the blue-heat temperature is sometimes termed "blue brittleness," though, as French and Tucker[110] pointed out, the reduced ductility is not accompanied by a corresponding drop in the impact curve. Figure 169, however, indicates that minimum impact values occur *near* the temperature at which maximum strength occurs. A maximum in the repeated-impact

curve for 0.07 and 0.08 per cent carbon steels was found at 210 to 220°C. (410 to 430°F.) by Kühle.[315]

Steel deformed in the blue-heat range and cooled to room temperature is found to have been hardened by such deformation to a greater extent than if the same degree of deformation had

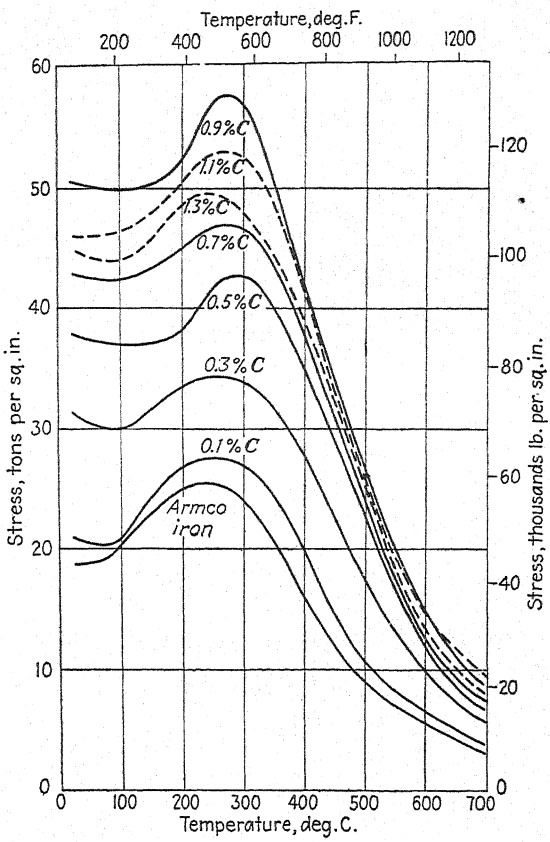

FIG. 168.—Tensile strength of Armco iron and annealed carbon steels at elevated temperatures. (*Inokuty.*[259])

been accomplished at room temperature. This may be illustrated by Sauveur's[408] results shown in Fig. 171.

As Sauveur,[408] Van Wert,[498] Kawai,[625] and others have shown, deformation at blue-heat temperatures in steels which are subject to age hardening is accompanied by a stepwise slip rather than by a uniform progressive slip, while in a non-aging

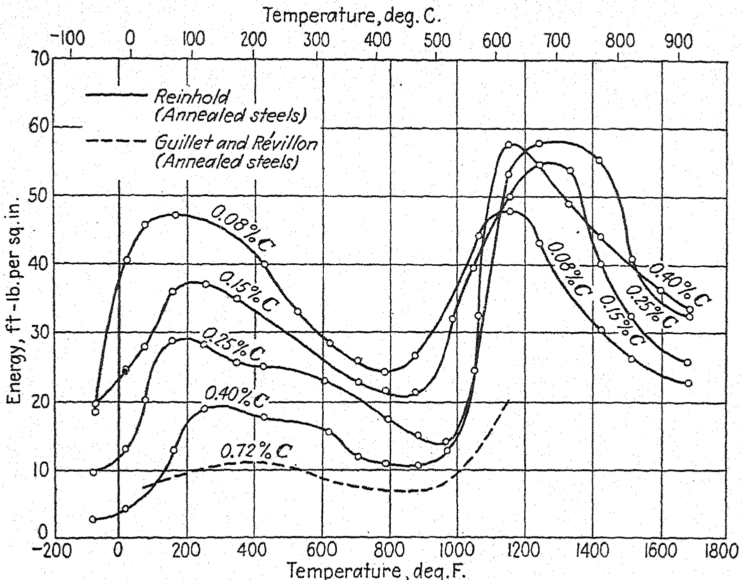

FIG. 169.—Notched-bar impact resistance of annealed carbon steels at elevated temperatures. (*Reinhold*[41] *and Guillet and Révillon.*[14])

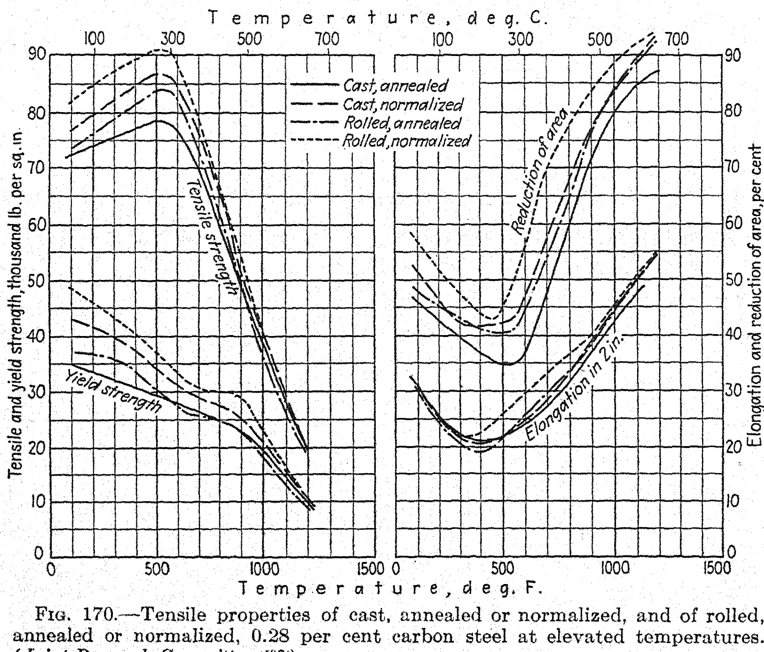

FIG. 170.—Tensile properties of cast, annealed or normalized, and of rolled, annealed or normalized, 0.28 per cent carbon steel at elevated temperatures. (*Joint Research Committee.*[583])

steel, like Izett, more nearly progressive slip occurs. Tentative explanations for the strengthening and for the mode of slip, advanced by Van Wert,[498] Kühle,[315] and Dean and coworkers,[293] are to the effect that some precipitation-hardening phenomenon is occurring in the steel at the temperatures in question. Mehl and Briggs in discussion of Van Wert's[498] paper, although holding that such an explanation may not be entirely adequate, stated that "it is the best explanation now available." Discussion has raged about the particular element, compound, or combination of these responsible for the effect, with carbon, oxygen, nitrogen, and manganese oxide among those prominently mentioned. It would appear that no single constituent is solely responsible, but more likely a combination of several, owing to the mutual interaction upon the solubility-temperature relationships.

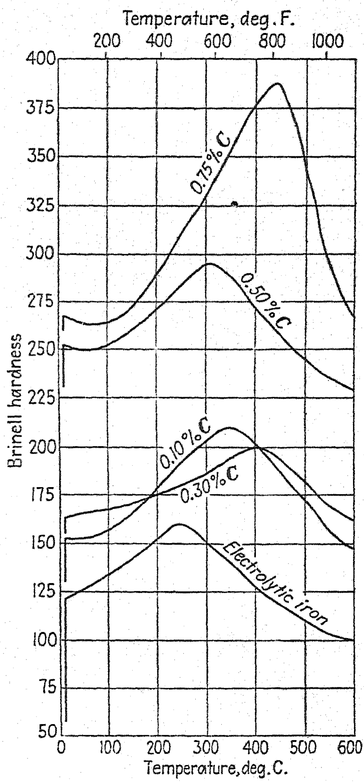

FIG. 171.—Brinell hardness at room temperature after straining electrolytic iron and carbon steels at the indicated temperatures. (*Sauveur.*[408])

Hayes and Griffis[711] as well as Kenyon and Burns[726] have shown on aluminum-treated steels, in line with the work of Van Wert[498] and others, that the addition of around 0.06 per cent titanium and 0.04 per cent aluminum to a steel of 0.05 per cent carbon, plus annealing at 640°C. (1180°F.) with slow cooling, practically eliminates both the aging propensity and, as is shown by Fig. 172, the rise in tensile strength in the blue-heat range. The titanium and aluminum might combine with both oxygen and nitrogen, rendering them inactive, so it is still doubtful which one is of primary importance in producing the precipitation hardening which causes the increase in tensile strength.

170. The Range of Secondary Brittleness.—The phenomenon of "secondary brittleness," observed around 450 to 600°C. (840 to 1110°F.), or around 450°C. (840°F.) for very slow application of load,[678] is of more than academic interest because of its possible, and indeed logical, connection with "shatter cracks" in steel rails, which are thought[368,559,560,566] to occur as the rail

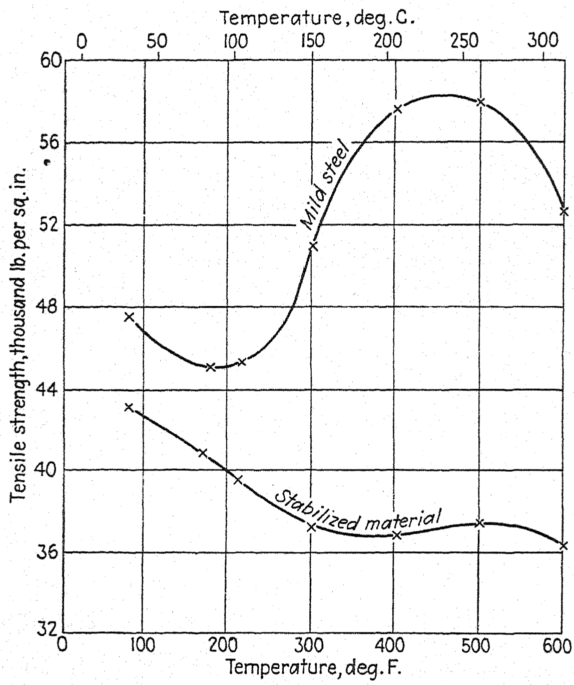

FIG. 172.—Difference in tensile properties, at elevated temperatures, of 0.05 per cent carbon-steel sheet and of the same sheet after stabilizing with titanium and aluminum. (*Hayes and Griffis.*[711])

cools on the hot bed. When the base and web have cooled to the blue-heat range, where the tensile and yield strengths are high, the interior of the head is just entering the secondary brittle range. Differential contraction, between the base and web, bends the rail and imposes tensile stress upon the head, and the difference in temperature between the outside and inside of the head also imposes a tensile stress. It is postulated that such stresses, acting on the interior of the head at a temperature at which it is both relatively weak in tension and lacking in ductility

(if the steel has a secondary brittle range), produce these shatter cracks, which have been shown to exist in new (unused) rails.[53] The cracks do not extend to the surface, nor have they been reported close to the end of a rail, indicating that they occur after the rail has been cut by the hot saw and that a certain degree of restraint must be exerted by the surrounding metal before the stresses build up sufficiently to cause cracks.

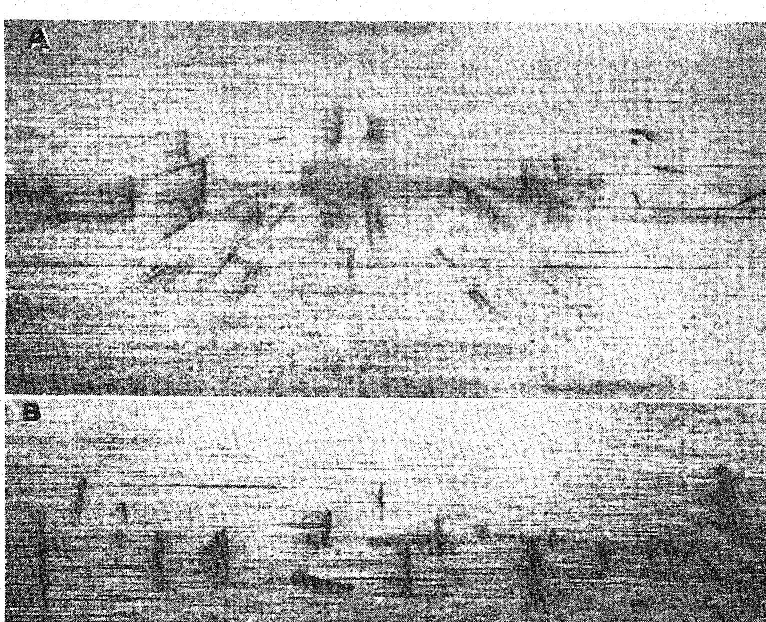

FIG. 173.—Shatter cracks in a new rail. A = horizontal longitudinal section; B = vertical longitudinal section. Natural size. (*Quick.*[559])

Not all rails show shatter cracks, but different lots of rails show marked differences in secondary brittleness. Figure 173 from Quick[559] shows shatter cracks in a new sorbitized rail containing 0.77 per cent carbon, 0.74 per cent manganese, and 0.30 per cent silicon. Figure 174, giving temperature-property curves of this steel, reveals a marked secondary brittle range.

Since the underlying causes of secondary brittleness are not known, so that steel-melting control cannot yet be exercised to avoid it, Freeman and Quick[368, 560] suggested that slow cooling through the range of secondary brittleness would be desirable for rails and similar sections in order to avoid shatter cracks, which

are thought by some to be nuclei of transverse fissures, so dangerous in rails. The Sandbergs(566) independently arrived at the same conclusion and devised "ovens" for slow cooling of rails through this range. Mackie claimed* that the temperature at which shatter cracks occur is lower than that indicated by the

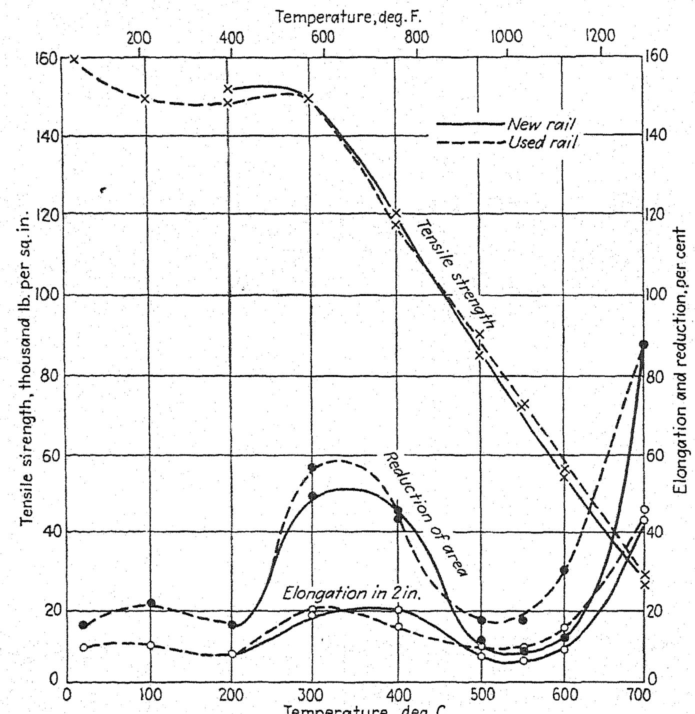

FIG. 174.—Temperature-property curves of sorbitized rail steel showing marked secondary brittleness. (*Quick.*(559))

work of Freeman and Quick and of the Sandbergs, and proposed to retard cooling to very low temperatures.

In discussion of another paper,(492) Quick gave curves showing that notched-bar impact tests, as do tensile tests, go through a range of secondary brittleness.

Secondary brittleness may be present in ingot iron, as was shown by Inokuty,(259) and in carbon steels of a variety of carbon contents as well as in various alloy steels, according to Rawdon

* Canadian Patent 319,553, Feb. 2, 1932.

and Berglund.[232] Annealing below the lower critical temperature sometimes appreciably decreases the degree of secondary brittleness, and this gives some indication that the phenomenon may be connected with some constituent, or constituents, of varying solubility at different temperatures which precipitate below the lower critical temperature. There is, however, little evidence to indicate just what constituents might be involved.

It is quite possible that, like the rise in strength in the blue-heat range and the variable shape and location of the notched-bar impact curve with temperature, secondary brittleness is not at all an inherent property of iron-carbon alloys but a result of adventitious impurities present in commercial iron and steel. Desch,[602] for example, suggested that it is due to nitrogen, oxygen, or carbon.

E. CAST FERROUS ALLOYS

The properties of cast steels at elevated temperatures are not radically different from those of wrought steels of the same composition. As shown by the data discussed below, in most cases the properties of wrought steels are slightly better at temperatures of 100 to 400°C. (210 to 750°F.) than the corresponding properties of cast steels; these differences tend to disappear at higher temperatures. In this connection Tapsell[496] remarked that between 400 and 500°C. (750 and 930°F.) carbon steel in the cast condition is probably no better, and may be worse, than annealed or normalized wrought steel.

In determining the difference in properties between a cast and a wrought steel of similar composition, investigators have generally used split heats, *i.e.*, one part is tested as cast and the other portion is hot worked to the desired section. Cast iron, as usual, is a law unto itself: the behavior of steel at elevated temperatures gives no clue to the properties of cast iron at corresponding temperatures.

171. General Relation of Tensile Properties to Temperature for Cast versus Wrought Carbon Steels.—Figure 170 (page 474) shows how closely the elevated-temperature properties of cast and wrought materials of the same composition agree. These results were obtained by three laboratories on a split heat containing 0.28 per cent carbon, 0.66 per cent manganese, and 0.35 per cent silicon. Both cast and rolled specimens were

annealed for 1 hr. at 870°C. (1600°F.) and normalized at 900°C. (1650°F.). The tensile strength of the rolled specimens at about 300°C. (570°F.) is some 5000 lb. per sq. in. higher than of the corresponding cast specimens; this difference becomes smaller until, at 650°C. (1200°F.), the tensile strength of all of the specimens is the same. The same general trends may be noted in the other properties, but the differences in elongation values are slight.

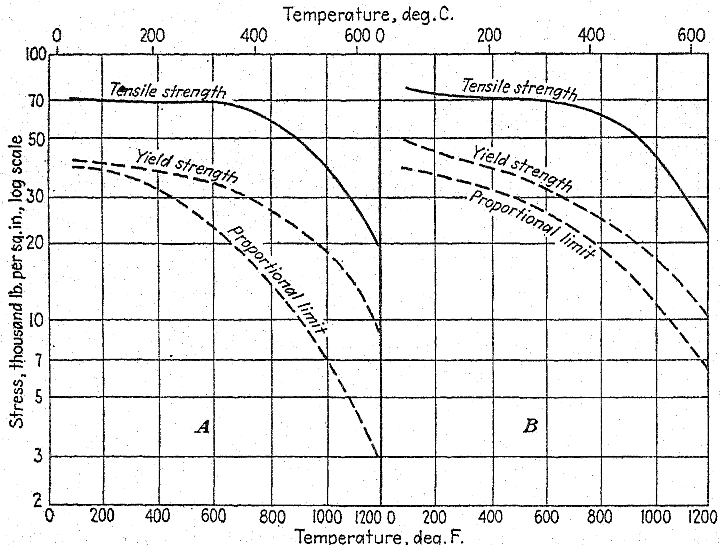

FIG. 175.—Elevated-temperature properties of *A* cast and annealed and *B* forged and annealed split heat, containing 0.20 per cent carbon. (*Spring*.[415])

Short-time tensile data from a split heat containing 0.20 per cent carbon are shown in Fig. 175.[415] Comparison of the curves indicates that the tensile and yield strengths for the cast and the forged material are about the same at all the temperatures used; the proportional limit of the forged specimens is higher for testing temperatures of 300°C. (570°F.) and above. Other short-time tensile tests were also made by Spring[415] on steels of similar composition but not from split heats. The data are plotted in Fig. 176. For the 0.35 per cent carbon steels the forged specimens had slightly higher properties, but for the 0.40 per cent carbon steels the properties of the forged specimens

were markedly superior at practically all the testing temperatures used. Creep properties were also determined on these steels;

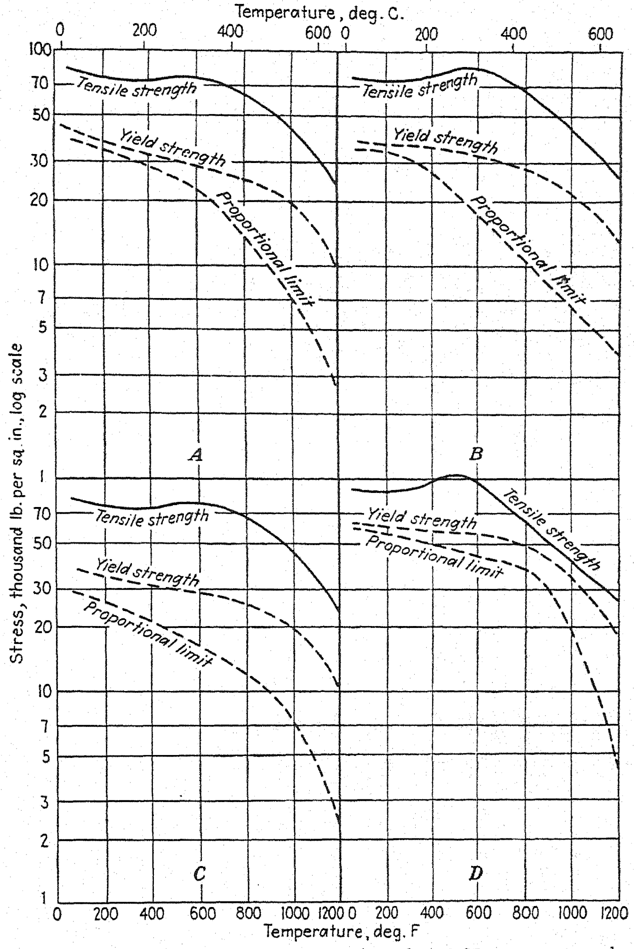

FIG. 176.—Elevated-temperature properties of *A* 0.33 per cent carbon steel, cast and annealed; *B* 0.35 per cent carbon steel, rolled and normalized; *C* 0.40 per cent carbon steel, cast and annealed; and *D* 0.41 per cent carbon steel, forged and annealed. (*Spring.*[415])

in general, the forged material had a higher creep strength than the corresponding cast specimens.

Bailey and Roberts[507] compared the short-time tensile properties of a cast 0.30 per cent carbon steel and a forged 0.40

per cent carbon steel; they found a tensile strength at 500°C. (930°F.) of:

Cast steel...........................	34,000 lb. per sq. in.
Forged steel.........................	51,500 lb. per sq. in.

Because of the difference in carbon content, nothing can be concluded from these two values.

Impact tests were made by Moffett[394] on cast carbon steels at temperatures between −40 and +20°C. (−40 and +70°F.). The tests were made at only three or four temperatures, and almost straight lines were drawn to represent impact resistance as a function of temperature; the familiar sharp drop in impact resistance was not found. Annealing the cast material improved the impact resistance at all testing temperatures used.

The differences in high-temperature properties between cast and wrought materials are, without doubt, largely due to the difference in grain size. This is brought out more clearly in the discussion of the creep properties of cast carbon steels in the next chapter.

172. Miscellaneous Data on the Effect of Temperature on Properties of Cast Carbon Steel.—Knipp[727] studied cast specimens of 0.20 per cent carbon steel (balance of composition not stated), which showed a rise in tensile strength at 250°C. (480°F.) and higher in tests carried out in 25 min. but showed a continuous decrease in tensile strength from room temperature to 400°C. (750°F.) when the specimen was broken in 30 sec. He found the knee in the curve, on slow testing, to occur at lower temperatures and to be the more marked the more rapidly the steel had been cooled from the austenite range. Low-temperature impact tests on cold-worked specimens that had been given an accelerated aging treatment were also made, and Knipp concluded that the change in tensile strength and the fall in impact resistance were both related to aging phenomena; he made the definite statement that the peak in the high-temperature strength curve is a result of precipitation hardening.

High-temperature fatigue data for a cast steel containing 0.53 per cent carbon were given by Tapsell and Clenshaw,[240] but the room-temperature endurance ratio of only 27.5 per cent instead of a normal one of at least 40 per cent raises some doubt whether the data are representative. The endurance tests were

carried out to only 10 million cycles, which is probably not sufficient for tests at temperatures as high as 500 or 600°C. (930 or 1110°F.).

In recent work Piwowarsky, Božić, and Söhnchen(649) determined the short-time tensile properties at temperatures of 650

TABLE 87.—COMPOSITION AND OTHER PERTINENT DATA OF THE MELTS STUDIED BY PIWOWARSKY, BOŽIĆ, AND SÖHNCHEN(649)

Steel	Composition, per cent										Strength becomes zero at		Reduction of area becomes zero at		Deoxidation with
	C	Si	Mn	P	S	N_2	O_2* (1)	O_2* (2)	SiO_2	Al_2O_3	°C.	°F.	°C.	°F.	
Electric steels															
A	0.26	0.49	0.76	0.054	0.013	0.004	0.0143	0.0171	0.010	0.025	1390	2535	1375	2505	†
B	0.22	0.35	0.78	0.063	0.025	0.004	0.0061	0.0141	0.007	0.022	1370	2500	1350	2460	
Basic open-hearth steels															
A	0.30	0.35	0.60	0.069	0.079	0.006	0.0038	0.0143	0.002	0.027	1400	2550	1350	2460	Ferrosilicon
B	0.34	0.34	0.62	0.084	0.080	0.006	0.0051	0.0155	0.002	0.022	1410	2570	1275	2325	
Acid open-hearth steels															
A	0.35	0.29	0.60	0.029	0.037	0.004	0.0088	0.0141	0.000	0.030	1360	2480	1270	2320	Ferrosilicon
B	0.41	0.33	0.73	0.025	0.028	0.005	0.0057	0.0196	0.010	0.035	1410	2570	1390	2535	
Bessemer steels															
A	0.18	0.23	0.77	0.102	0.134	0.006	0.0180	0.0277	0.025	0.036	1460	2660	1410	2570	0.02 per cent Al
B	0.28	0.51	1.26	0.100	0.085	.....	0.0070	......	.....	.....	1440	2625	1410	2570	

* (1) Vacuum fusion; (2) chemical extraction.

† Deoxidation with ferromanganese and ferrosilicon; 0.5 kg. per ton aluminum in the ladle.

to 1450°C. (1200 to 2640°F.) of cast steels containing between 0.2 and 0.4 per cent carbon. The steels were made by different processes (electric furnace, Bessemer, and open-hearth), and the properties at high temperatures were found to be dependent on the steel-making process.

The primary object of the investigation was to determine the cause of hot-cracks in cast steel. Rectangular slabs were cast

upright, chilled on two sides to produce transcrystallization, and specimens were taken with and across the direction of the crystals, on which short-time high-temperature tests were made. Only tensile strength and reduction of area were recorded. The test bar was radiused to a minimum at the mid-section and had no cylindrical gage length, apparently being thus chosen, because of poor temperature distribution along the bar, in order to make sure that the temperature recorded by the thermocouples at the

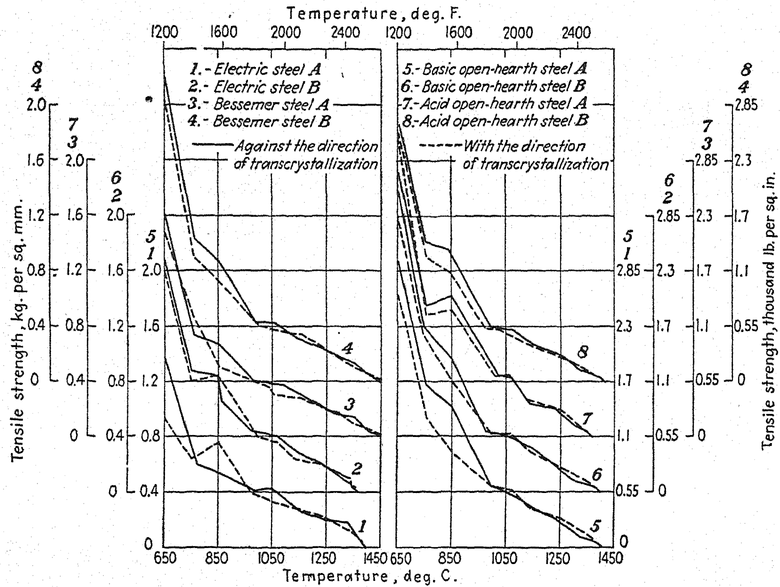

FIG. 177.—Effect of elevated temperatures on the tensile strength of cast carbon steels made by different processes. See Table 87 for composition and details of deoxidation. (*Piwowarsky, Božić, and Söhnchen.*[649])

minimum section represented that of the active portion of the test bar. The core of the bar was calculated to vary in temperature from that at the locations of the thermocouples by as much as 7°C. (13°F.). The tests are therefore of a special nature, representing neither an ordinary sand casting nor an ordinary high-temperature tensile test.

The compositions and mode of deoxidation are given in Table 87.

The results are shown in Figs. 177 and 178. On account of incipient melting (analogous to the "burning" or overheating of

steel) at around 1350°C. (2460°F.) the reduction of area fell to zero.

No appreciable difference was noted in the tensile strength of all the steels above 1000°C. (1830°F.). Notable differences were

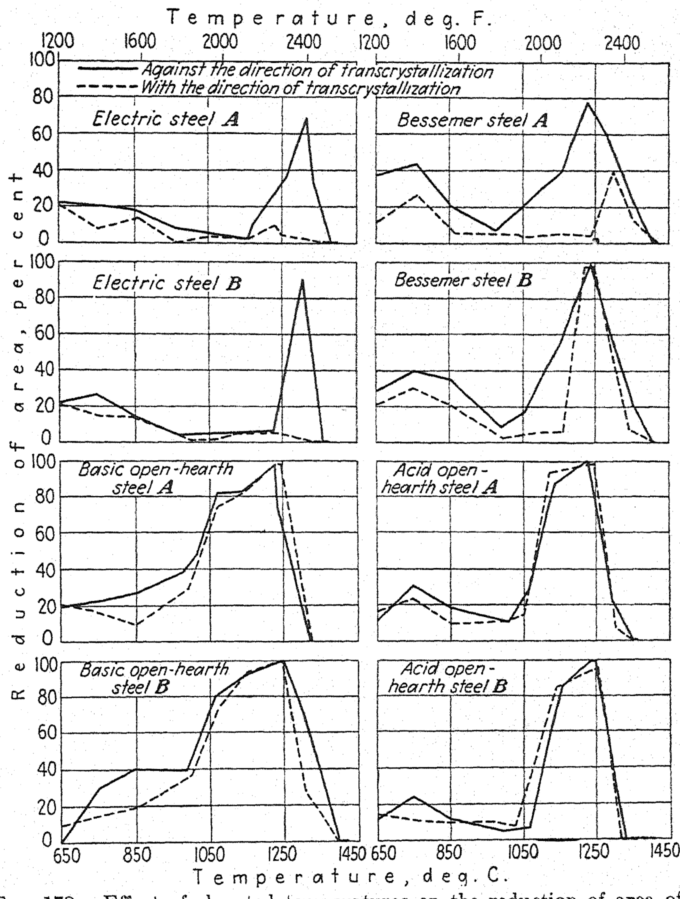

FIG. 178.—Effect of elevated temperatures on the reduction of area of cast carbon steels made by different processes. See Table 87 for composition and details of deoxidation. (*Piwowarsky, Božić, and Söhnchen.*[649])

found in reduction of area at temperatures around 1050°C. (1920°F.), especially in the electric- and Bessemer-steel specimens taken in different directions in relation to the direction of the crystals. These steels all had aluminum added, while the acid and basic open-hearth steels did not.

A detailed study of the results obtained by Piwowarsky, Božić, and Söhnchen should be of interest in relation to the work of Sims and Lillieqvist(571) on inclusions in cast steel.

173. Effect of Temperature on the Properties of Cast and Malleable Irons.—As stated on page 292, the modulus of elasticity of cast iron is indeterminate, the stress-strain curve being curved from the origin even at room temperature. The slope of the curve

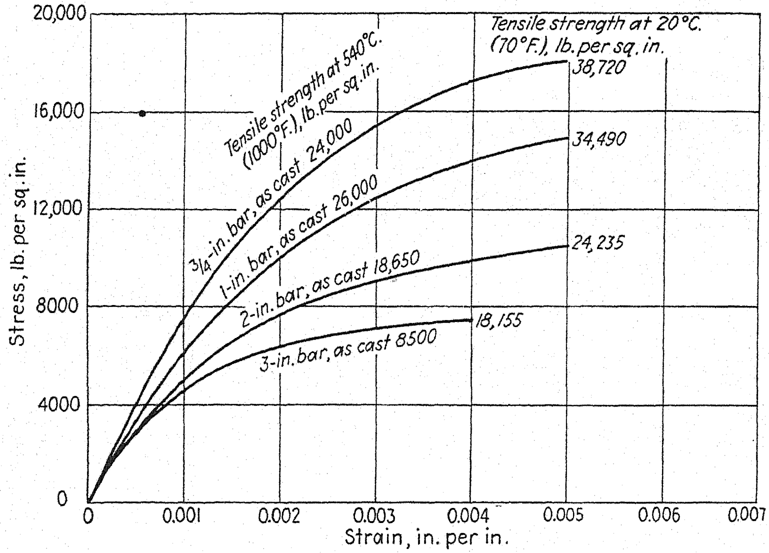

FIG. 179.—Effect of section size on the stress-strain curves at 540°C. (1000°F.) of a gray cast iron containing 3.25 per cent carbon and 1.70 per cent silicon. (*Bolton and Bornstein.*(437))

also varies with the size of the casting as is indicated by Fig. 179 from Bolton and Bornstein(437) which shows stress-strain curves at 540°C. (1000°F.) for a cast iron containing 3.25 per cent carbon and 1.70 per cent silicon. These investigators also reported data on the tensile properties of four gray irons at temperatures up to 540°C. (1000°F.). The properties were determined on 0.505-in. specimens, machined from cast 1.25-in. bars for irons Nos. 1, 2, and 3 tested at 425°C. (800°F.) or below; and machined from cast 1.0-in. bars for all of the tests at 540°C. (1000°F.) and for the tests of iron No. 4 at 425°C. or below. The data are given in Table 88.

TABLE 88.—SHORT-TIME TENSILE STRENGTH OF FOUR CAST IRONS*

Iron number	Composition, per cent		Tensile strength, lb. per sq. in., for a testing temperature of			
	Total carbon	Silicon	20°C. (70°F.)	230°C. (450°F.)	400°C. (750°F.)	540°C. (1000°F.)
1	3.54	2.17	22,000	17,000	20,000	12,000
2	3.51	1.71	27,000	28,000	28,000	20,000
3	3.31	1.63	32,000	33,000	34,000	24,000
4	2.90	1.75	42,000	39,000	41,000	26,000

* Bolton and Bornstein.[437]

In commenting on these and other properties reported in the same paper, Bolton and Bornstein stated:

> Gray irons, irrespective of classes or grades, when judged by short-time tests, appear to have a very slight change from their original tensile strength between room temperature and 425°C. (800°F.). In certain cases [for example, irons Nos. 1 and 4, Table 88] a slight increase may appear between 315 and 425°C. (600 and 800°F.). Above 480°C. (900°F.) there is a sharp lowering in strength in all grades, which lowering continues with increase in temperature.

These investigators also gave data which indicate that the tensile strength, elongation, and reduction of area of malleable iron change very little between room temperature and about 425°C. (800°F.). Above this temperature the tensile strength decreases and the elongation and reduction of area increase.

Results of high-temperature tensile and endurance tests, reported by Moore and Lyon[225] for a cast iron of 3.44 per cent total carbon, 0.68 per cent combined carbon, 1.10 per cent silicon, 0.62 per cent manganese, 0.51 per cent phosphorus, and 0.09 per cent sulphur, are shown in Fig. 180. The iron was cast as hollow cylinders, 18 in. long, 12 in. inside diameter, and 14 in. outside diameter. In the "prolonged-loading" tensile tests each test lasted "several days."

Endurance tests at elevated temperature on a gray cast iron were also reported by Kaufmann.[465] The specimens were taken from hollow cylinders, 27.5 in. high, 25.5 in. inside diam-

eter, and with a wall thickness of 1.6 in. The iron contained 3.19 per cent total carbon, 1.09 per cent silicon, 0.82 per cent manganese, 0.12 per cent phosphorus, and 0.13 per cent sulphur.

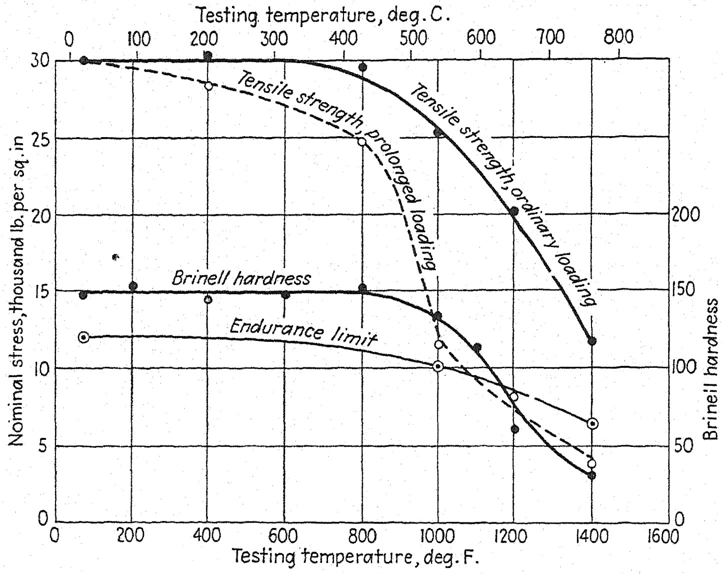

FIG. 180.—Elevated-temperature properties of gray cast iron containing 3.44 per cent total carbon, 0.68 per cent combined carbon, and 1.10 per cent silicon. (*Moore and Lyon.*[225])

Its tensile strength at room temperature was 38,000 lb. per sq. in. The endurance limits found were:

Temperature		Endurance limit, lb. per sq. in.
°C.	°F.	
20	70	14,000
100	210	13,000
200	390	13,000
300	570	15,500
400	750	16,500
500	930	14,000
600	1110	Less than 10,000

So few tests were run at each temperature that these values are uncertain.

It will be noted that neither the ordinary short-time loading nor the more-prolonged-loading tensile test indicates that cast iron has increased strength in the blue-heat range, such as is met in ordinary steel. A similar absence of a region of strength

TABLE 89.—COMPOSITION AND ROOM-TEMPERATURE PROPERTIES OF CAST IRONS WHOSE HIGH-TEMPERATURE PROPERTIES ARE SHOWN IN FIG. 181*

Kind of iron	Composition, per cent						
	Total carbon	Graphite	Combined carbon	Silicon	Manganese	Phosphorus	Sulphur
Gray..........	3.40	2.70	0.70	1.50	0.75	0.15	0.03
High test......	2.95	2.10	0.85	2.45	0.75	0.05	0.10

Room-temperature properties

Kind of iron	Brinell hardness	Transverse tests, 1.2-in. diameter, 12-in. centers		Set under proof load,† in. per in.	Endurance limit, lb. per sq. in.	Charpy impact,‡ ft-lb.
		Breaking load, lb.	Deflection, in.			
Gray..............	197	3400	0.15	0.0024	14,500	6.5
High test..........	237	4500	0.14	0.0001	22,500	6.1

* Allen.(428)

† Proof load of 25,000 lb. per sq. in. applied 10 min., load released, and permanent set measured.

‡ Large specimens, 4.06 in. long, 0.80 in. square, 3.20-in. span, V-notch 0.16 in. deep.

higher than that at room temperature is shown in the compression tests made by Sale.(763) Sale used cast iron of 3.52 per cent total carbon, 0.42 per cent combined carbon, 1.58 per cent silicon, 0.37 per cent manganese, 0.71 per cent phosphorus, and 0.12 per cent sulphur, cast in hollow rounds, 1.5 in. outer diameter, 0.5 in. inner diameter. The compression specimens had an

effective length of 9.25 in. The compressive strengths obtained were as follows:

Temperature °C.	Temperature °F.	Compressive strength, lb. per sq. in.
Room temperature		87,000
250	480	80,000
310	590	66,000*
410	770	72,000
515	960	49,500
635	1175	23,000
755	1390	7,850

* Irregularities in the stress-strain curve indicate that this value is not reliable and Sale's temperature versus compressive-strength curve is accordingly dotted at this point.

The curve for these values is slightly above that for structural

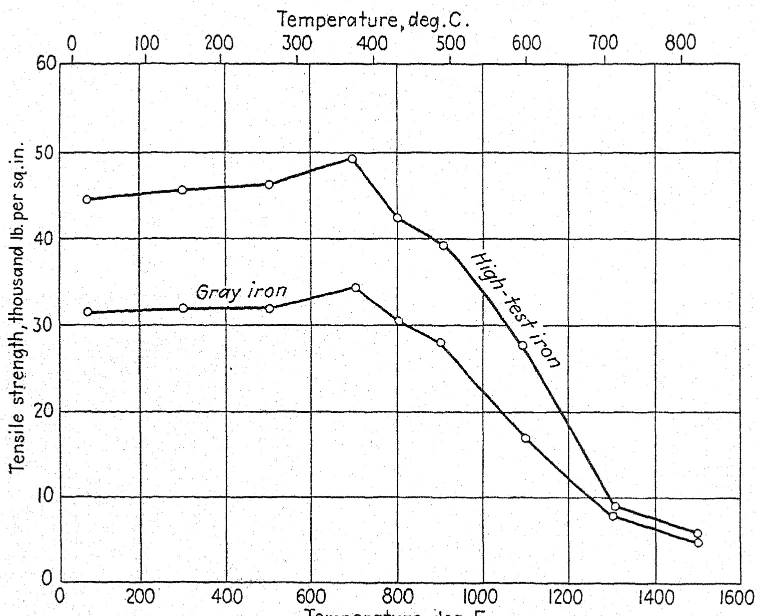

FIG. 181.—Elevated-temperature tensile strength of gray cast iron and high-test cast iron. See Table 89 for composition. (*Allen*[428])

steel at all temperatures. Figure 181, from a discussion by Allen,[428] shows that there may be a slight hump for high-test

cast iron. The composition and room-temperature properties are shown in Table 89.

F. AUTHOR'S SUMMARY

1. Within the last decade many data on the effect of temperature on mechanical properties of ferrous alloys have been reported. Carbon steels lose strength so rapidly at elevated temperatures that alloy steels and special heat-resisting alloys are used if service conditions are severe; hence, most of these data were obtained on alloy steels rather than on carbon steels. As a result, despite the large amount of work which has been done, many fundamental facts about the high- and low-temperature properties of carbon steels cannot yet be stated definitely.

2. In general the hardness and strength of a steel vary inversely with the temperature, while the ductility varies directly. Not only does the rate of change in properties vary with the temperature, but in some steels actual maxima or minima are observed in curves representing a property as a function of temperature. It is, therefore, necessary to determine properties not only over a wide temperature range, but also over rather small temperature intervals. The modulus of elasticity decreases with increasing temperature; it is, however, difficult to determine the modulus or to locate exactly the proportional limit at temperatures above 300 or 400°C. (570 or 750°F.).

3. With the exception of notched-bar impact resistance, the rate of change of properties at temperatures slightly above or below room temperature is small. At temperatures slightly below normal, the properties, except in the case of impact resistance, change more gradually with change in temperature than at still lower temperatures; the notched-bar impact resistance of many steels, however, drops precipitously near room temperature. The exact temperature at which this large drop occurs depends upon the composition and the previous treatment of the material. The true effect of carbon on the properties in the impact-embrittlement range cannot yet be evaluated.

4. In general, values for endurance limit at elevated temperatures follow the same trend as the values for tensile strength. Data are given which indicate that the endurance ratio of carbon steels may be substantially constant for temperatures from normal to about 500°C. (930°F.). There are not enough data

available to show that this constancy of ratio holds for all carbon steels and all conditions of testing. The data do indicate, however, that up to about 400°C. (750°F.) the endurance limit of low- and medium-carbon steels is at least the same as, and at some intermediate temperatures even higher than, at room temperature when this limit is obtained by rapidly reversed stresses alternating about a zero mean value. When the stress is not completely reversed, or when the reversal is slow, distortion by creep may occur. Oxidation, unless prevented, may also affect the results. The few data available on endurance at subnormal temperatures indicate that the endurance limit rises proportionately with the tensile strength.

5. In alternate-bending tests which, according to Schwinning, Knoch, and Uhlemann,[769] give values 3 per cent higher than rotating-beam endurance tests, carbon steels containing 0.12 to 0.60 per cent carbon show a peak at 300 to 400°C. (570 to 750°F.) which is 100 to 200°C. (180 to 360°F.) higher than the peak in tensile strength for the corresponding steel. A noteworthy phenomenon in these tests is that the alternate-bending strength for 10^5 reversals is as high as, or higher than, the tensile strength for the same steel at temperatures of 400°C. (750°F.) or above. Whether this phenomenon holds for all carbon steels or only for steels of the composition and structural condition used by these investigators should be determined by further work. The alternate-bending strength for these steels at 300°C. (570°F.) and above is far greater than the yield strength. A 45-deg. V-notch reduces greatly the alternate-bending strength; the amount of reduction depends upon the depth and sharpness of the notch.

6. At subnormal temperatures tensile and yield strengths increase slightly with fall in temperature, and reduction of area decreases. The limited data available indicate that elongation is affected but slightly. The few data quoted also indicate that the modulus of elasticity of medium-carbon steel is slightly lower at low temperatures than at room temperature, but for high-carbon steel the modulus is unchanged.

7. As noted above, the impact resistance of many carbon steels decreases abruptly at temperatures which may be but slightly below normal. Marked differences in impact resistance at these slightly subnormal temperatures may be apparent in steels of approximately the same composition (as usually deter-

mined). Much of the variation in low-temperature impact resistance in low- and medium-carbon steels of apparently the same composition is due to a difference in grain size. Herty and McBride[717] have shown that fine-grained steels have better low-temperature impact resistance than coarse-grained material. It is even possible that grain size exerts such a marked influence on low-temperature impact resistance that the alleged specific effects of small amounts of alloying elements are in reality outweighed by differences in the degree of deoxidation of the material.

8. The work on the relation between low-temperature impact resistance and grain size has shown clearly that no definite conclusions on the effect of impurities or of small amounts of manganese, silicon, chromium, copper, and other elements (with the possible exception of nickel) can be formulated unless the grain-size variable is first taken into consideration. Although it is probable that carbon steels of small grain size have better low-temperature impact resistance than corresponding coarse-grained steels, a high degree of low-temperature toughness is apparently impossible to attain in carbon steels as a class; hence, in those applications where shock must be withstood at very low temperatures, alloy steels are usually specified.

9. Between room temperature and the blue-heat temperature —about 300°C. (570°F.)—the strength, hardness, and endurance limit rise to a peak, while elongation, reduction of area, and notched-bar impact resistance fall. After the blue-heat temperature has been passed, the tensile strength, yield strength, hardness, and endurance limit fall and, in general, the ductility and notch toughness increase. At still higher temperatures, say above 500°C. (930°F.), plastic flow becomes very evident. The peak in strength and in hardness at the blue-heat temperature may not be a true phenomenon of the iron-carbon system. As some non-aging steels do not show this peak, it is possible that it may be caused by adventitious impurities.

10. In the neighborhood of 450 to 600°C. (840 to 1110°F.) many steels show "secondary brittleness," in other words, minimum ductility values. It is possible that this phenomenon may also be caused by impurities and that secondary brittleness might not be observed in *pure* iron-carbon alloys. The available data indicate that shatter cracks in rails form in the range of secondary brittleness.

11. The properties of cast carbon steels at elevated temperature do not differ greatly from the properties of wrought steels of the same composition. In most cases, the properties of the latter are slightly superior to those of cast steels in the temperature range of 100 to 400°C. (210 to 750°F.), but the superiority tends to disappear at higher temperatures. It seems likely that such differences as exist in the elevated-temperature properties of cast and wrought carbon steels of the same composition are largely the result of differences in grain size.

12. The tensile strength of gray cast iron, regardless of class or grade, apparently changes very little in the range between room temperature and 425°C. (800°F.). A possible exception is high-test cast iron; data are quoted which show a slight peak in the tensile-strength curve at temperatures of 370 to 400°C. (700 to 750°F.). Short-time tests of gray cast iron do not show an increased strength in the blue-heat range, as is characteristic of commercial carbon steels. Above about 480°C. (900°F.), there is a sharp lowering of the tensile strength with increasing temperature. Malleable iron, also, changes but little in tensile strength, elongation, and reduction of area at temperatures below about 425°C. (800°F.). Above this temperature, the strength decreases, and the elongation and reduction of area increase.

CHAPTER XIII

EFFECT OF TEMPERATURE ON PROPERTIES: II. CREEP TESTS

Creep and Creep Testing—Creep Data on Wrought Carbon Steels—Creep Data on Cast Ferrous Alloys—Author's Summary

In many types of service at elevated temperatures steel fails by distortion beyond the amount permitted for the particular application in question or by fracture, at stresses far below those of the short-time, high-temperature tensile- and yield-strength values or even below the proportional limit. The increasing loads and temperatures being used in modern steam-power plants and in oil cracking, for example, call for exact evaluation of the resistance of steels to slow deformation at high temperatures and for the development of alloy steels more suitable than carbon steels for severe service.

A. CREEP AND CREEP TESTING

As mentioned in the previous chapter, the technique of testing at temperatures above atmospheric is difficult, the equipment costly, and the work tedious. This is true for short-time tests but even more true for creep testing. At ordinary temperatures, the rate of application of load in a tensile test is of relatively minor importance. At elevated temperatures it is of such major importance that it may be said quite accurately that high-temperature properties of any material cannot be fully evaluated by a rapidly conducted tensile test. The determination of the rate of deformation produced by a given load over a period of several hundreds or even thousands of hours is required to secure the data not given by the short-time tensile test.

Because of the importance of an understanding of what goes on during plastic deformation, and of the methods used to determine it, a discussion of these two factors prefaces the correlation and summary of the actual creep data of carbon steels given later in this chapter.

174. Distortion of Carbon Steels under Load at Room Temperature.—Few quantitative data are available on the distortion under long-continued load of carbon steels at room temperature, since strain hardening of a steel stressed at a constant load slightly beyond the elastic limit soon stops the deformation. Thurston,[1] quoted by Howe,[2] made some early tests on annealed and unannealed irons which showed that, when loaded at 90 and 80 per cent of its short-time tensile strength, annealed iron broke respectively in 35 and 91 days, but at 70 per cent or less it was unbroken after 17 months, while unannealed iron broke in 266 and 455 days at 80 and 70 per cent load respectively but was unbroken after 17 months at 60 per cent. The initial deformation followed by strain hardening, characteristic of creep curves, was shown in these early tests by Howe for annealed copper. Moisseiff[395] studied patented cold-drawn bridge wire with 0.75 to 0.80 per cent carbon of 220,000 to 235,000 lb. per sq. in. tensile strength and 103,000 lb. per sq. in. proportional limit. The wire was stressed, at room temperature, to 113,000 lb. per sq. in. The extension found occurred mostly in the first two weeks, was practically complete after three weeks, and at the end of two months amounted to 0.00024 in. per in.

175. Plastic Flow and the So-called "Amorphous-metal" Hypothesis.—In addition to the temperature ranges discussed in some detail in the previous chapter, there is another dividing line of temperature, not very clearly marked, above which "creep" becomes of greatest importance. (Creep is a property only qualitatively indicated in the short-time tests and so faintly indicated by them that an entirely different testing technique must be adopted to reveal the facts.) This dividing line is the softening temperature at which hardening, from prior cold deformation, is removed. The softening temperature is in the general range of what has been called the "equicohesive-temperature range."

At low temperature, steel is elastic for low loads (it follows Hooke's law) and only begins to show plastic deformation at the elastic limit. Cast iron, however, has some plastic behavior at room temperature; the stress-strain curves for soft irons are curved from the origin. If steel is stressed above the elastic limit, plastic deformation, *i.e.*, slip, sets in, becoming marked at and above the yield point. In the range between the elastic

limit and the yield point, deformation is partly elastic and partly plastic. For metals like lead, for which room temperature is a relatively high temperature in comparison with the melting point, one is accustomed to see the plastic effect overbalance the elastic effect, and one expects such metals to creep under low loads; lead roofing and lead piping are thus designed accordingly.

For each metal and alloy there is some temperature range below which the behavior for low loads is essentially elastic and above which it is essentially plastic. This temperature range roughly coincides with the softening temperature, the temperature at which the effect of cold work is removed, but since the softening or annealing action is the resultant of time and temperature, the temperature is not a very definite one.

It is recognized that plastic materials like glass and pitch may be very brittle at low temperatures and that at medium temperatures they may appear rigid on rapid loading but will slowly flow or sag on long-time loading. At a given temperature a pitch may be "brittle" and yet in time flow out of a small orifice in a barrel. From the analogy of slow flow of amorphous materials like pitch, there was built up, a couple of decades ago, the idea that metals contain an amorphous constituent. It was obvious that the crystalline grains could not be amorphous; so it was postulated that a submicroscopic boundary layer of "cement" about the crystal is amorphous. The shift from elastic to plastic behavior at some temperature range was then ascribed to the softening of the amorphous cement; this view was made plausible by the fact that at low temperatures, when the cement was thought to be stiff, fracture takes place *through* the crystals, and at higher temperatures, when the cement would naturally be weak, it takes place *between* the crystals. The temperature at which this shift occurs was termed the equicohesive temperature.

While the more cautious early advocates of the amorphous theory were careful to phrase their comments to the effect that the boundary material *acted* as though it were amorphous, others were less hesitant and fell into calling it actually amorphous. Several studies of the past decade indicated that the boundary material is essentially crystalline and not "without orientation," as a strict definition of the term amorphous would require. Layers of sputtered metal only a few atoms thick

have been prepared which appeared structureless when examined by X-rays but which shifted over to crystalline form on very slight heating. Recent work, however, indicates that the advocates of amorphous metal are in a defensible position.*

Opinions differ on how much slightly imperfect or strained crystal lattices approach the behavior of a pitch-like material, and some advocates of the amorphous theory have altered the definition of amorphous to include the strained or deformed lattice. The equicohesive temperature was formerly defined as that at which the amorphous material is weaker than the crystalline and was said to mark the change from intracrystalline to intercrystalline fracture. This change in type of fracture more or less coincides with the softening temperature of cold-worked materials, and hence with the end of strain hardening by deformation, as well as with the "relaxation" temperature described by Scott.[490]

While it is true that one may expect the intercrystalline type of fracture at high temperatures and the intracrystalline type at low temperatures, the dividing line is not sharp. The best dividing line from the point of view of creep is the end of the strain-hardening range, for below this range there is some assurance that the rate of creep shown after a few hundred hours will decrease owing to strain hardening, even in spite of some decrease in cross-section, while above it the decrease in cross-section is not compensated for by strain hardening, and it is possible that the rate of creep will ultimately increase. Thus, there is some semblance of logic in using the term "creep limit" for the resistance to creep below the end of the strain-hardening range, but it is inapplicable above it.

176. The So-called Creep Limit.—The allowable deformation of a material for steam-turbine design may be less than 0.01 per cent per yr. over a period of 25 years and, while test methods are not yet adequate to evaluate accurately such low rates of deformation and to allow reliable extrapolation to such periods from those of actual testing, engineers are desirous of having data of that degree of precision. Obviously, the short-time tensile values are no direct measure of such slow yielding as is involved;

* See, for example, the articles of E. Rupp, *Kolloid Zeitschrift*, v. 69, 1934, pp. 369–376, and of G. P. Thomsen, *Nature*, v. 135, 1935, pp. 492–495.

hence, special creep tests must be made in order to evaluate the resistance to deformation under continued loading. To determine even the relatively huge deformation, from the point of view of creep, of 1 per cent in 1000 hr. requires rather precise measurements of elongation, exact control of temperature, and a long period of test.

Many of the older data on creep were obtained under test conditions such that the precise control of variables, judged by present standards to be essential, was not attained and often not approached. These data are usually wholly unreliable for design purposes and are useful only as rough indications, though much useful pioneer work was done with relatively crude equipment and methods. Unfortunately, owing to lack of understanding at that time of the importance of some of the variables involved, early published reports often failed to include information on test methods essential to a correct appraisal of the validity of the results. The situation is further complicated by the opinion formerly held by some that there is a true creep limit at any temperature, which opinion has led to many statements in the literature regarding values for creep limit or "load for no creep," when what was actually meant was a final rate of creep that seemed very small in the eyes of the investigator at the time, or for his own purpose, or which represented merely the limit of sensitivity of the equipment then used. Then, too, many statements regarding behavior at "elevated temperature" are too broad and sweeping, because they ignore the fact that there are two distinct regions above room temperature in which steels act differently, *i.e.*, the region below and the region above the softening temperature.

Steel work-hardens on being deformed, *i.e.*, it is cold worked even though the temperature is above that of normal cold-working processes, and, if the temperature is not so high that the effect of cold work is removed by annealing as fast as it is produced, the steel which is deformed may still be capable of sustaining the load without risk of eventual failure. Hence, the design load, below the softening temperature, may be high if initial deformation can be permitted, or it may be relatively low if the criterion of failure is deviation from *initial* dimensions. In this range a creep limit will depend entirely on the definition of allowable degree of deformation.

Above the strain-hardening range, there is increasing doubt whether there is any load that will not actually cause some minute deformation, and engineers who deal with this upper temperature range are beginning to consider it so and to design for materials that continuously stretch. The question then becomes whether the *rate* of creep will be too great for the purpose in hand over the period of service expected. Creep is, therefore, being expressed, not so often as the stress for a given creep limit where the criterion must be arbitrarily defined before it means anything, but as stress for a given rate of creep, usually in terms such as 1 per cent in 10,000 hr. or 0.1 per cent in 10,000 hr., etc. Days, months, or years are sometimes used for the time, but the common units are 10,000 or 100,000 hr. Even where strain hardening is not involved, the initial rate of creep is normally much greater than that which occurs later, so that it is necessary to differentiate between total creep, which includes the initial deformation, and final rate of creep, which excludes it. Most of the data now being reported represent the final rate of creep. It would be helpful to have the information presented in both ways.

177. Problems in Creep Testing.—In all studies of plastic flow the rate of loading is important. High rates of loading indicate higher resistance of the material than do low rates. Similarly, the rate of extension under a given load at a given temperature is not constant in the early stages of extension. Even in the region above the strain-hardening range the initial rate is greater than the rate hundreds or thousands of hours later. When creep has progressed so far that necking down occurs, the rate again increases till final failure ensues. The problem might then be phrased as whether this inversion in rate can occur, from the extrapolated data, within the desired life of the structure. Many attempts have been made to use methods of testing carried out somewhat more slowly than the short-time tensile test but still only carried on for a few days, and—by comparing the rate of extension from the "3d to the 5th hour" or from the "24th to the 72d hour"—to predict what the rate will be after thousands of hours.

In order to compare the high-temperature properties of steels for use in high-pressure superheated steam lines, a Subcommittee of the American Society for Testing Materials is collecting the

available creep data and is planning to take into account only such data as have been obtained with the precautions laid down in the code, and to disregard any creep tests which have not been carried to at least 1000 hr. The rate of creep between the 700th and 1000th hr. is to be used as one of the factors in the evaluation of the properties. This insistence upon data based on long-duration tests, by engineers who have given much thought both to methods of making the tests and to the industrial application of the data, is indicative of the necessity for a reasonably sound basis for the extrapolation that is still required.

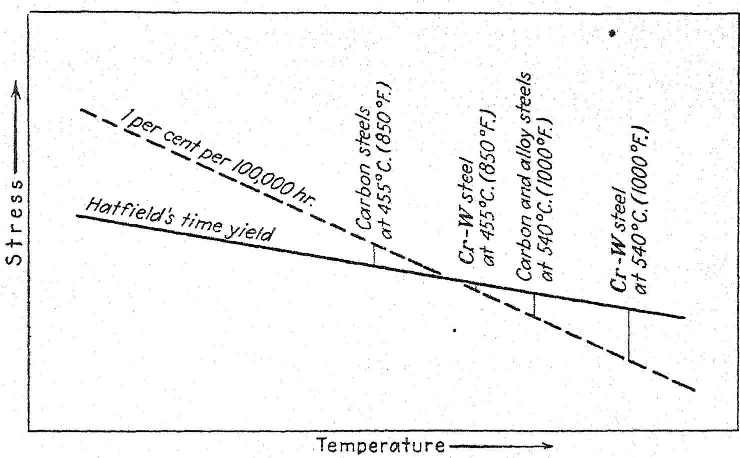

FIG. 182.—Diagram of the relation between Hatfield's time-yield and the stress for a creep rate of 1 per cent in 100,000 hr. at various temperatures. (*White and Clark.*[783])

Foley, in discussion of a paper by White and Clark,[784] showed creep curves which indicate that in some cases it may take 2000 hr. or more for even the general trend of the creep curves to become evident.

178. Accelerated Creep Tests.—White and Clark[783] examined several types of steel at a range of temperatures and found, as shown diagrammatically in Fig. 182, that the relation between creep and one of the most widely heralded accelerated methods (Hatfield's[376] time-yield test) varied with both composition and temperature. They found that the time-yield test "does not give values which can be directly converted into actual creep data," though they considered that the method has "possi-

bilities" as a relatively rapid, but only qualitative, method of study.

Pomp and coworkers,[557,755] in investigations of the accelerated method for determining creep developed at the Kaiser-Wilhelm Institut für Eisenforschung,[230,402] made a study in which assumed values for loads to give a final creep rate of 1 per cent in 10,000 hr., guessed at from the rate of extension between the 5th and 10th hr., were compared with actual creep curves. (Few of the latter, however, were carried on for more than 500 hr., all too short a period for reliable results.)

Pomp and Höger[557] made comparisons at 400 and 500°C. (750 and 930°F.). At 400°C. (750°F.) the correlation was considered by them to be satisfactory save in the case of a quenched and tempered nickel-chromium steel for which they assumed, from the accelerated test, that a load of 31,000 lb. per sq. in. would give a final rate of 1 per cent in 10,000 hr. but which, actually so loaded, gave a creep rate of about 4 per cent. There was also a considerable discrepancy at 500°C. (930°F.). In other tests at 500°C. the value for a steel of 0.11 per cent carbon was guessed to be 6700 lb. per sq. in., but their creep data indicated only 5700; and a nickel steel of 0.18 per cent carbon and 1.56 per cent nickel, for which a value of 25,000 lb. per sq. in. had been assumed, gave a higher final rate of creep than 1 per cent in 10,000 hr. when loaded at 24,000 lb. per sq. in. Nevertheless Pomp and Höger concluded that, except for heat-treated steels, their method was useful.

More recently Pomp and Herzog[755] made comparisons at 550 and 600°C. (1020 and 1110°F.). It seldom took more than 100 hr. of an actual creep test to show that the accelerated test led to values far too high, although a few 500-hr. and a couple of 800-hr. tests were made. A steel of 0.37 per cent carbon for which the assumed value was 3100 lb. per sq. in. actually showed less than 1400 lb. per sq. in. at 550°C. (1020°F.), while another of 0.04 per cent carbon and 0.83 per cent copper with an assumed value of 13,000 lb. per sq. in. actually showed but 3600. At 600°C. (1110°F.), only alloy steels were studied. Two which were assumed to have a value of 14,000 lb. per sq. in. each actually showed 5000 and 3500, while two others, both actually showing less than 1400 lb. per sq. in., had been assigned values of 6400 and 4300 lb. per sq. in. for a rate of 1 per cent in 10,000 hr.

Pomp and Herzog agreed with other investigators that such accelerated methods have no useful function for temperatures above 500°C. (930°F.), *i.e.*, above the strain-hardening range. It logically follows that for any given steel the annealing temperature which no longer allows strain hardening during creep must be experimentally established and that even at lower temperatures the behavior in the early stages of creep needs to be experimentally correlated with that in the later stages. Hence the answer, obtainable only by long-time creep tests, must be known before the accelerated-test results can be utilized.

Thus it appears that the only possible advantage of early-stage tests is as acceptance tests to find out if different lots of steel of the same composition and thermal history do act exactly alike in creep. It often happens, however, that similar specimens behave quite differently in the early stages of creep, but after 500 or 1000 hr. their creep curves will coincide, so that early-stage acceptance tests might lead to the rejection of entirely satisfactory material.

179. Methods Based on Variable Load or Temperature.—"Relaxation tests" have been tried in which a specimen is first highly loaded and creeps, and then either the load[509] or the temperature[590] is reduced until creep stops, within the sensitivity of the apparatus. As Bailey and coworkers[433] stated, such methods are "on trial" only. They involve huge extrapolations and, while they may be useful in preliminary studies such as are contemplated by Austin with the equipment described by Austin and Gier[590] (see discussion), data obtained by them are not yet generally considered acceptable or reliable substitutes for actual creep data. Many of the data in the literature obtained by such methods must be discarded for the present at least.

Ranque and Henry[757] improved the accuracy of Rohn's[405] accelerated test (constant load and variable temperature) by refinements in technique. However, they did not provide for an adequate time factor. They showed a curve based on 14 observations made in 4 days on two machines, using four test specimens, at temperatures ranging from 200 to 750°C. (390 to 1380°F.), which is alleged to establish the creep limit for 5 per cent flow in 1000 hr. (a rather large rate of flow compared to the 1 per cent in 10,000 or 100,000 hr. desired by most American engineers). However, they failed to give any true creep tests by which the

reliability of their figures might be appraised. It is not clear whether these authors view the Rohn test as one for preliminary appraisal of the high-temperature properties of a group of alloys in which it may have a legitimate use, or as a means of securing data applicable to problems of engineering design. They probably take the former view since they stated that the method is particularly valuable for laboratory investigations. Lacking any comparison with long-time creep or service data, it would appear that they might well have warned the engineer against trusting such data until a sound relationship with long-time load-carrying capacity has been established.

McVetty[740] has proposed a method by which, when sufficient long-duration creep data are available and assurance of metallurgical stability is given, extrapolation may be made on the basis of correlation of the values into families of curves consistent with the known behavior of many types of steels. He presented such families of curves for 0.30 per cent carbon cast steel and 0.40 per cent carbon wrought steel.

Loading a given specimen to a high stress, establishing its rate of flow, and then lowering the stress upon the same already stretched material and again measuring the rate of creep has been used,[514] but most investigators believe this "step-down" practice tends to give high results. Coffin and Swisher[514] said: "The behavior of a material at high temperature depends upon the manner of loading, and, therefore, the numerical values of creep strength have significance only when viewed in the light of procedure and apparatus used in making the tests."

The "step-up" method, in which the specimen, at a given temperature, is first loaded so low that it does not stretch at all, or after about 1000 hr. maintains a constant or decreasing rate of flow, and in which the stress is then raised for another such period, is used more often than the step-down method and seems less objectionable. But Kanter and Spring[380] produced evidence to show that this method also is "inaccurate."

180. Conclusions on Creep Testing.—For the most valuable comparison of creep characteristics it seems necessary to use but one load and one temperature for each specimen and to carry the test on for a long period. The tentative creep code proposed by the Joint Research Committee on the Effect of Temperature on the Properties of Metals[583] requires that tests to be reported

in terms of load to produce a creep rate of 0.1 per cent in 10,000 hr. shall be carried on at least 1000 hr., and those of 1 per cent in 10,000 hr. for at least 500 hr. It also provides for limits of temperature variation, for uniformity of temperature over the gage length of the specimen, and for sensitivity of elongation measurements, and it provides that individual specimens be used for each load and temperature. These requirements all appear essential to accuracy, as was forcibly brought out in discussions by Gillett and Cross[614] and by Foley.[610] Relatively few of the many data in the literature have been obtained under the conditions specified in the code, and many of the relationships that should be made clear on the effect of different variables on creep resistance are obscured by the inaccuracies and inconsistencies of the older data on creep.

Dustin[690] reported that the Belgian Research Committee on Behavior of Metals at Elevated Temperature, utilizing the results of work in other countries, studied the various high-temperature testing methods which had been recommended and, both from published data and from direct experiment, decided that none of the accelerated methods was reliable and that tests of over 1000 hr. were required. "Prototype" steels are selected by evaluation by the long-time tests, and acceptance tests are expected to be made on the basis of agreement of several creep curves for the specimens under study with those of the prototype steel during tests of say 130 hr. duration.

Clark and Robinson[793] have reported creep tests on low-alloy steels which were carried out for 10,000 hr., and which raise interesting questions on the extrapolation of shorter time tests, but which unfortunately were not so carried out as to answer them.

181. Importance of Creep Data.—Pending the realization of the hope of development of trustworthy rapid methods for determination of long-time load-carrying ability—a hope still far from fulfilment—designing engineers are asking more and more emphatically for definite data on long-time creep resistance of steels at elevated temperatures. Obtaining the needed data is both time consuming and expensive, but engineers are slowly realizing that the information is worth its cost.

Until the fundamental laws of creep become much more clearly apparent, so that valid generalizations and some correlation with

other more readily determinable properties can perhaps be made, it will be necessary that the problem continue to be attacked on the basis of individual instances before suitable information will be available for evaluation of the mutual effect of many variables occurring together. Hence, for an answer to most specific questions of engineering design for high-temperature service, special creep tests need to be made on the specific materials and at the specific temperatures under consideration.

That both engineers and metallurgists are becoming "creep conscious" is evidenced by a casual comparison of the papers in the 1924[(110)] and the 1931[(431)] joint A.S.T.M. and A.S.M.E. symposia on Effect of Temperature on Metals and by perusal of Tapsell's[(496)] 285-page book, which is entirely devoted to a discussion of creep of metals.

Nevertheless, even the most reliable laboratory creep data fail to throw direct light on many engineering questions that are of an equal order of importance. The situation is rather analogous to that of the determination of the endurance of metals under repeated stress, where much effort was required before the details of test specimens and test methods were so worked out that consistent results were obtainable by different laboratories and where much time was spent in chasing the mirage of "accelerated methods." After endurance-test methods came into use which did give results reproducible in any laboratory and masses of data were accumulated and correlated, it became evident that the endurance limit was but one of the highly important criteria in judging the suitability of materials for service under repeated stress, and that intelligent engineering judgment in evaluating the various properties of the materials in the light of the particular service could not be replaced by the single value for endurance limit.

Similarly, a definite knowledge of creep rates of materials for high-temperature use, based on 1000-hr. laboratory tests for instance, does not tell the whole engineering story. Large extrapolations are still necessary to predict behavior over many years of service; the short-time high-temperature and impact properties need to be determined and considered when occasional overloads or shocks are imposed in service; and most important of all the question must be answered whether the material under-

goes slow structural changes during long exposure at the temperatures of service.

Unless the material is initially in a stable state, changes, such as in microstructure or those of a precipitation-hardening or agglomeration-softening nature, may occur which are not clearly brought out even in several thousand hours, still less in but a few hundred hours, of creep testing. The problem is not so difficult in the case of carbon steels as it is in the case of some of the newer alloy steels, for the structural stability of carbon steels and the factors affecting it are better understood. The precipitation-hardening effects in carbon steels, however, are not yet so well clarified.

The range of temperatures over which carbon steels are usable is smaller than for the heat-resistant alloys, so that less work is required to establish the behavior in the important range.

For all the steels there is some range of commercially important temperatures in which the properties change rapidly. Unless the actual temperatures of service are known with exactness—and the engineer's knowledge of temperature fluctuations in service is generally very hazy—a perfectly correct choice of materials for the *alleged* temperature, based on creep and other data at that temperature, may be an erroneous one for the *actual* temperature range. As Swift put it in a discussion:[509] "Safe working stress depends on the uncertainties as to the maximum stress and maximum temperature in service and can never, therefore, be determined purely by tests of materials at high temperatures." The effect of oxidation and the corrosive effect of the particular atmosphere surrounding the material in service and at service temperatures introduce disturbing factors which are often of the highest importance.

Thus knowledge of creep resistance alone tells but a small part of the story. That part, nevertheless, is a most essential one, and its determination under conditions such that the propensity toward creep is separated from the other variables and evaluated by itself is indispensable for the engineer's complete appraisal of the situation. Some engineers are temporarily blinded by their excessive interest in creep data and neglect the other important factors. A few are equally blinded by their abhorrence for any test method so tedious and expensive as creep, refuse to

obtain or utilize creep data, and will consider nothing but service data or those obtained under "fully simulated" service conditions, in spite of the difficulty of describing or duplicating exactly all service conditions with their manifold variables and the need for such large amounts of data that statistical methods can be used for evaluation. The open-eyed middle course will, of course, be the one of progress.

Some of these points of view are interestingly brought out in a discussion of a paper by Eaton(606) in which MacQuigg made the sensible proposal that practical limits for the definition of creep and method of determining it be agreed upon for immediate engineering use, that study of short-time tests for rapid-inspection purposes be continued, and that research laboratories continue the laborious searching for fundamentals.

The promulgation of the codes(583) for short-time and for creep-test methods is an important step along the path outlined.

B. CREEP DATA FOR WROUGHT CARBON STEELS

Reliable quantitative data on carbon steels are largely lacking, not only because of the unstandardized or non-standard test methods formerly in vogue but also because the creep resistance of carbon steels is so low at high temperatures that most attention has been paid to the "heat-resisting" alloy steels of better creep resistance.

It would be very helpful if all laboratories reporting creep tests on alloy steels would make available at the same time creep data on carbon steels of approximately the same basic analysis. Bailey and associates(433) pointed out that designers know the service behavior of medium-carbon steel at 400°C. (750°F.) and that the divergent values of creep reported by different investigators might be more nearly reconciled if their values for high temperatures were looked at as percentages of what they find at 400°C. (750°F.).

182. General Trend of Creep Characteristics.—Marked differences are shown in curves for creep (strain) versus time at relatively low temperatures, where strain hardening can occur, compared with those for higher temperatures, where it cannot occur. Such families of curves were shown by French and Tucker.(153) The smoothed curves are utilized to show the relations diagrammatically in Fig. 183 from Tapsell.(496) The

curves 1 to 7 are in the order of decreasing stress. Comparing these, it is seen that at high stresses, such as from 1 to 5, the specimen will ultimately neck down and break, since the rate passes through an inflection, and that, to have reasonable assurance that at lower stresses more creep than is permissible will not occur before the time (calculated from the final rate of creep) when the test is stopped (*e.g.* at 40 days or 1000 hr.), it is necessary to have very small rates of deformation, as in 6 or 7. While these curves were not scaled, curve 6 will probably correspond to a creep rate of 1 per cent in 10,000 hr. and 7 to 0.1 per cent in 10,000 hr. Where strain hardening can occur,

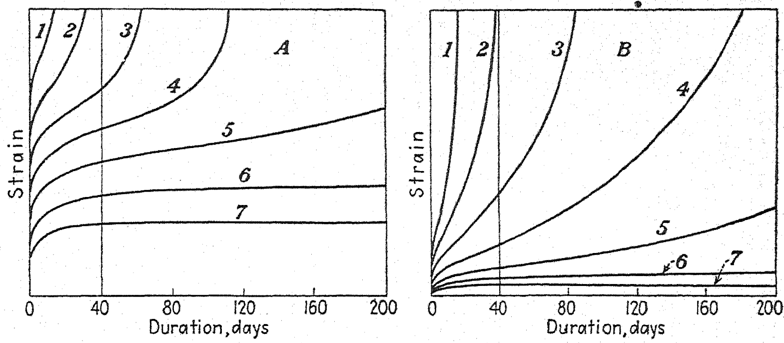

FIG. 183.—Diagrammatic representation of time-strain curves for a steel at a temperature where strain-hardening is *A* considerable and *B* small. Curves 1 to 7 are in the order of decreasing stress. See text above. (*Tapsell.*[496])

the initial deformation corresponding to a small *final* rate of deformation is much greater than where it cannot.

Figure 184, from Kanter and Spring,[260] shows the tensile properties at temperatures up to 650°C. (1200°F.), and the stresses that give several creep rates at different temperatures, of a 0.17 per cent carbon steel. The creep rates give *total* creep and, therefore, the extension of the specimen during the first 10,000 hr. In Fig. 185, from a discussion by Kanter,[433] double logarithmic plotting is used to indicate the rate of total creep of a 0.33 per cent carbon steel at different temperatures and at different loads. It will be seen that the relations found by Kanter for temperatures between 400 and 540°C. (750 and 1000°F.) agree rather well with the relationship shown by Bailey and coworkers[433] in Fig. 186 which indicates the creep resistance at temperatures above 400°C. (750°F.) as a percentage of the

creep strength at that temperature. Bailey's curve, however, refers to a minimum rate of creep instead of total creep.

183. Effect of Carbon Content on Creep Characteristics.—At temperatures up to 425°C. (800°F.) increase in carbon content, in plain carbon steels at least up to 0.6 per cent, raises the creep

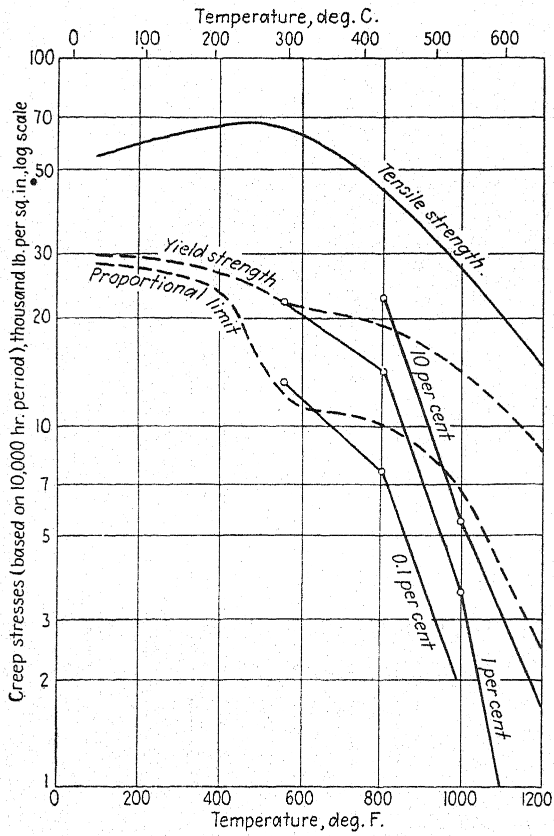

FIG. 184.—Stress-temperature curves for long-time and short-time tensile tests of a rolled and normalized basic open-hearth steel containing 0.17 per cent carbon. The stress ordinate is logarithmic. (*Kanter and Spring.*[260])

resistance, but at higher temperatures the increase brought about by increased carbon content is small or undetectable. This was commented on by French in discussion of a paper by Spooner and Foley[492] and was shown diagrammatically by Tapsell.[496] Some of the data used by Tapsell are shown in

Fig. 187; the so-called "creep limit" is the stress which produces a rate of elongation of about 4 per cent in 10,000 hr. at the end of 1000-hr. tests. The steels had been normalized.

In a later appraisal, Tapsell(663) stated that in annealed or normalized wrought carbon steels the creep resistance of a 0.10 per cent carbon steel is about 10 per cent below, and that of

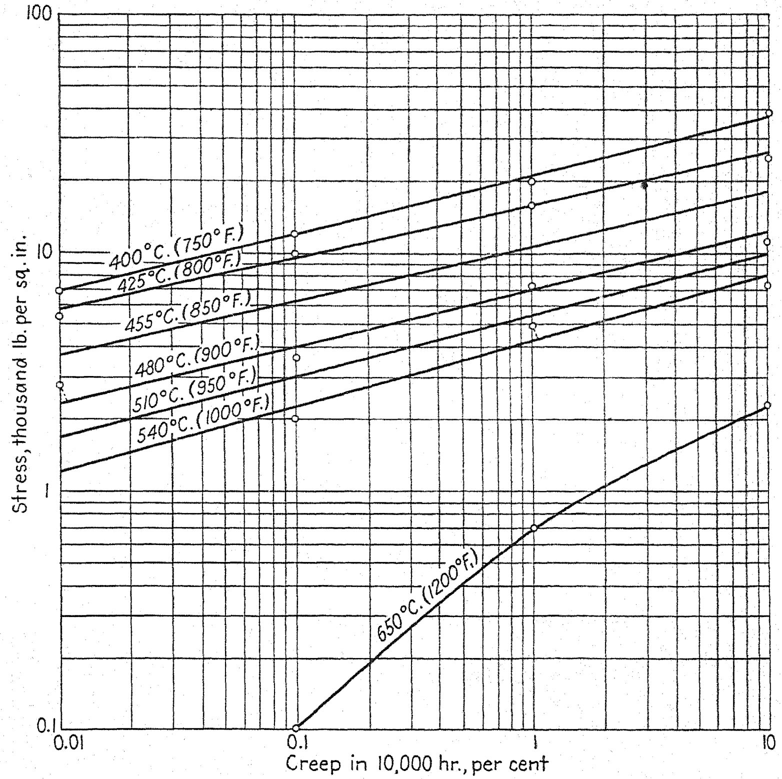

FIG. 185.—Relation of stress to total creep for a 0.33 per cent carbon steel. (*Kanter, in discussion of paper by Bailey and coworkers.*(433))

a 0.50 per cent carbon steel about 20 per cent above, that of a 0.25 per cent carbon steel taken as a base-line. He remarked, however, that the creep resistance of cast steel does not necessarily increase with carbon content.

The creep properties of steels containing high carbon, eutectoid or hypereutectoid steels, apparently have not been studied except for the investigation of Pomp and Dahmen,(230) who

included a spheroidized 1 per cent carbon steel in early "accelerated" creep tests which covered only the initial stage of deformation under load and gave no acceptable values for true long-time load-carrying ability. The rough tests made indicated that under the conditions of the test such a steel deformed under much lower loads at 400 to 500°C. (750 to 930°F.) than did much lower carbon steels containing lamellar pearlite.

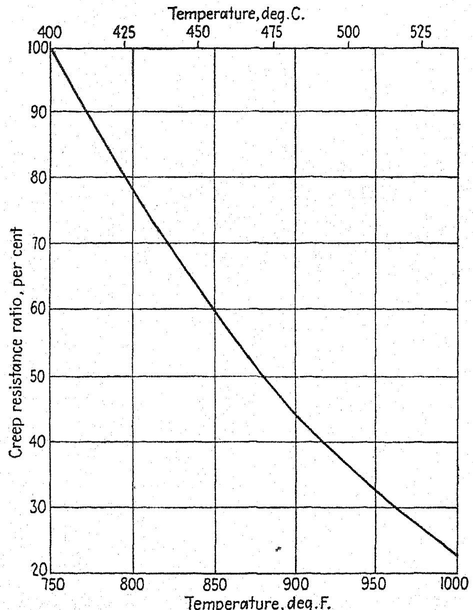

FIG. 186.—Creep resistance of medium-carbon steels plotted as percentage of the resistance at 400°C. (750°F.). (*Bailey and coworkers.*[433])

It was found by Kanter and Spring[260] that at 425°C. (800°F.) an open-hearth iron and a Swedish iron flowed more rapidly under a stress of 10,000 lb. per sq. in. than did steels containing from 0.17 to 0.40 per cent carbon; but at 540°C. (1000°F.) and a stress of 3000 lb. per sq. in. there was no appreciable difference, while at 650°C. (1200°F.) and a stress of 500 lb. per sq. in. the open-hearth iron flowed somewhat less than a steel containing 0.35 per cent carbon.

Clark, in discussing a paper by Bull,[443] reported the results of creep tests on a 0.13 per cent carbon steel made "by alloy-steel practice" which indicated that the stress required to give a rate

of deformation of 1 per cent in 10,000 hr. was 13,000 lb. per sq. in. at 425°C. (800°F.) and 11,000 lb. per sq. in. at 480°C. (900°F.); in a "commercial" 0.18 per cent carbon steel similarly tested the tests indicated that a stress of only 7000 lb. per sq. in.

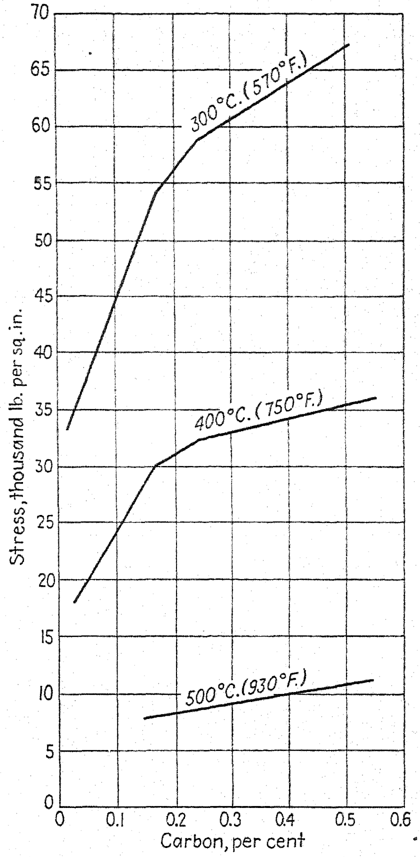

FIG. 187.—Relation between carbon content and the stress required to produce a rate of deformation of about 4 per cent in 10,000 hr. The steels had been normalized. (*Tapsell.*[496])

at 455°C. (850°F.) produced the same rate of flow. According to a discussion by Kinzel,[492] a silicon-killed steel has a higher creep resistance than a semikilled or rimmed steel. Whitney[499] remarked, with respect to tubing, that not only the composition but the steel-making and tube-making processes alter the high-

temperature properties. Hence, one can hardly expect that the recorded creep data for carbon steels will be consistent unless these variables are also recorded.

Dauvergne[683] made tests of four carbon steels at different loads and temperatures for periods up to 36 hr., in order to obtain a figure which he termed "elastic limit," defined by him as the load which will produce an extension of 0.0005 per cent per hr. in the interval between the 25th and 35th hour. Since no long-time creep tests were recorded, the relation of Dauvergne's data to actual creep properties is not clear. His curves, however, show differences in the same direction as those reported by Clark and by Kinzel between rimming and killed steels. The steels were made in the *basic* open hearth and were of the following composition:

Steel	Element, per cent					
	C	Si	S	P	Mn	Cu
A, rimming steel	0.09	Trace	0.023	0.019	0.54	0.08
B, boiler-tube quality	0.12	0.06	0.027	0.013	0.62	0.07
C, high-quality	0.12	0.11	0.025	0.014	0.61	0.09
D, high-silicon	0.12	0.19	0.024	0.010	0.62	0.08

Analysis for aluminum was made, but only faint traces were found.

TABLE 90.—LONGITUDINAL TENSILE AND IMPACT PROPERTIES AND HIGH-TEMPERATURE "ELASTIC LIMIT" OF RIMMING AND KILLED LOW-CARBON STEELS*

Steel	Tensile and impact properties					Load for elongation of 0.0005 per cent per hr. between the 25th and 35th hour, lb. per sq. in., at		
	Tensile strength, lb. per sq. in.	Yield strength, lb. per sq. in.	Elongation, per cent	Reduction of area, per cent	Mesnager impact, kg.	400°C. (750°F.)	450°C. (840°F.)	475°C. (890°F.)
A	49,000	35,500	37	72	31	11,400	7,400	5,100
B	55,500	42,500	35	74	29	13,000	9,800	6,500
C	57,500	44,000	35	65	28	20,000	17,000	10,500
D	59,500	43,500	32	59	15	16,500	12,700	9,000

* Dauvergne.[683]

The tensile and impact properties and the so-called "elastic limit" as determined by Dauvergne's method are given in Table 90.

Dauvergne concluded that the 0.11 per cent silicon steel was correctly killed and that the one with 0.19 per cent silicon con-

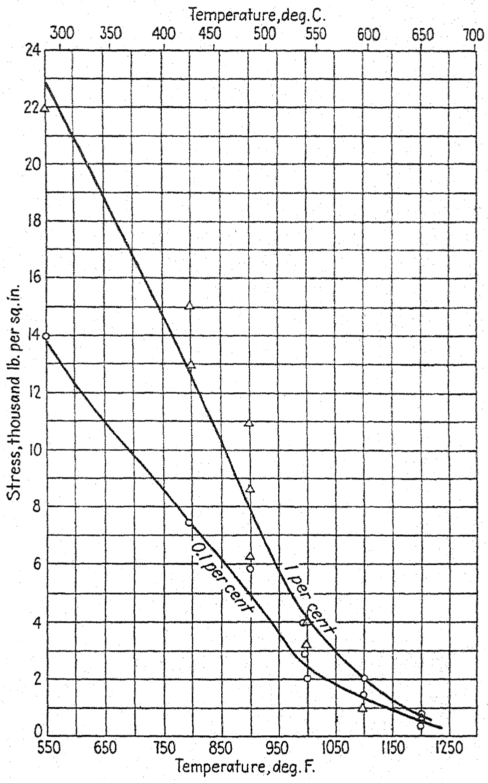

FIG. 188.—Stresses for creep rates of 0.1 per cent and 1 per cent in 10,000 hr. for steels containing 0.08 to 0.17 per cent carbon.

tained inclusions which perhaps accounted for its lower impact value. No data on grain size were given.

184. Correlation of Creep Data on Carbon Steels.—Data on creep of carbon steels obtained by long-time methods and reported in terms corresponding to 0.1 or 1 per cent in 10,000 hr., as obtained by Spooner and Foley,[492] Norton,[328] French and coworkers,[67,153,212,452] Kanter and Spring,[260] and Spring,[415] and some unpublished results obtained by a user of carbon-steel

tubing, have been collected and separated into groups of 0.08 to 0.17 per cent carbon, 0.18 to 0.35 per cent carbon, and 0.40 to 0.45 per cent carbon. These data, plotted in Figs. 188, 189, and 190, show a wide scatter when there is a fairly large number of points.

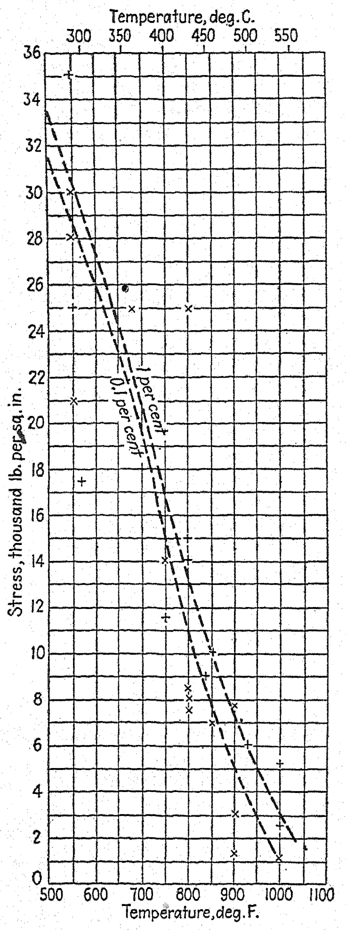

FIG. 189.—Same as Fig. 188, for steels containing 0.18 to 0.35 per cent carbon.

The central solid line in Fig. 191 shows average values of stress for a rate of deformation of 0.1 per cent in 10,000 hr. for all of the steels, and the shaded area shows about the range of scatter of the data. The bulk of the data giving the lower boundary of the band probably refers to total creep, while those giving the upper boundary quite certainly refer only to the final rate of creep. Design stress values for oil-still tubes of carbon steel as compiled by Dixon[448] are shown as circles in Fig. 191. The crosses are the design values given by Jasper in discussion of a paper by Kanter and Spring,[260] and the triangles are those given by Jacobus.[379] The design values for 650 and 705°C. (1200 and 1300°F.) may appear high, but they are for tubes that may bulge without serious consequences. The values are probably near the curve representing the stresses that will produce a deformation of 5 per cent in 10,000 hr. In testing, it is now common practice to determine the stress that will produce a rate of deformation of 0.1 or 1 per cent in 10,000 hr., but for purposes of design a higher or a lower rate may be important.

While the creep situation was complicated by the lack of standardization of test methods, prior to the promulgation of

the creep code of the Joint Research Committee on the Effect of Temperature on the Properties of Metals,[583] it nevertheless appears that the scatter of published data is not due solely to discrepancies in testing methods, but that a considerable part of the variation must be caused by inherent differences in the steel itself, differences which cannot be traced to the composition,

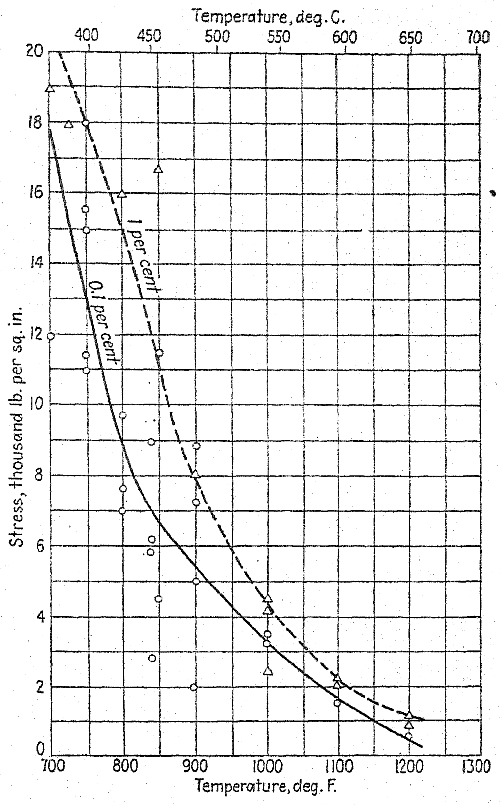

FIG. 190.—Same as Fig. 188, for steels containing 0.40 to 0.45 per cent carbon.

at least in the elements usually determined. It is recognized that for the best creep resistance the steel should be in an initially stable state, but, except for a very few studies of split heats, little information is available, based on the same heat of steel, showing the properties of cast versus wrought, annealed versus wrought, annealed versus normalized, normalized versus normalized and tempered material, etc.

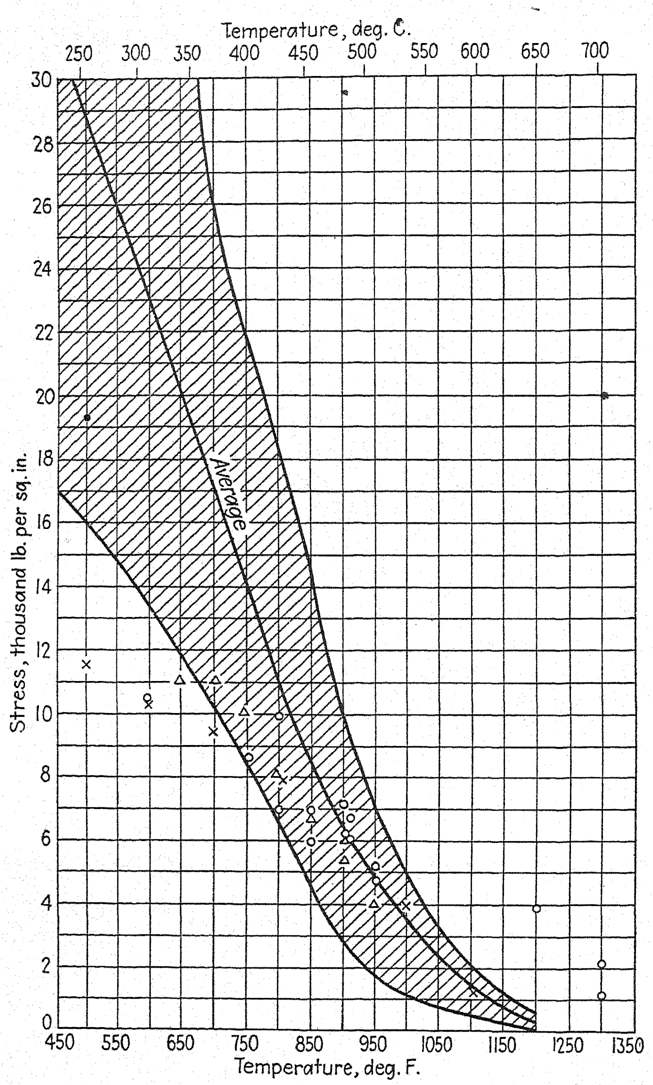

FIG. 191.—Creep properties of carbon steels. The central solid line is the average for a rate of deformation of 0.1 per cent in 10,000 hr. for all the values used for Figs. 188, 189, and 190, and the shaded area is the range of scatter. The plotting points are design values; see text.

Creep properties of carbon steels appear to be predictable only within limits that are far too broad for the satisfaction of the designer who has to use them at high temperatures.

The standard creep tests have been made with tensile stresses, but Everett, in discussion,[433] reported torsional-creep data for a carbon steel. For pure shear stress, as in torsion, according to Bailey and associates,[433] the stress required to produce a given rate of deformation is about half of that required to produce the same rate in tension. Curves were given by these writers to show the shear stresses at different temperatures required to give different rates of flow in steel containing from 0.2 to 0.3 per cent carbon.

185. Absence of Direct Relationship between Proportional Limit and Resistance to Creep.—It will be noted that short-time proportional-limit values are not generally reproduced here and are not compared with creep properties. It was hoped by some of the workers on high-temperature properties of metals, in earlier days, that useful data in regard to long-time load-carrying ability at high temperatures might be obtained by sufficiently precise determinations of proportional limit. Clark and White[290] pointed out that, while the proportional limit may be of the same order of magnitude as the load for long life, at temperatures above the equicohesive temperature the correlation is not necessarily good. However, they considered the proportional-limit values useful if obtained by sufficiently refined apparatus. It is now well established that the load for long life falls below the proportional-limit curve at some temperature, say 450 to 650°C. (840 to 1200°F.), which temperature varies with the particular steel tested. Or, expressed in another manner, the proportional-limit versus temperature curve is not parallel to the curve showing the stress required to produce a given amount of deformation in a given time, and no simple relationship between the two curves is evident.

The short-time tensile and yield strengths are considered of value in showing the ability of the metal to sustain a short overload, and tentative methods for their determination have been codified by the Joint Committee on the Effect of Temperature on the Properties of Metals,[583] but it is stated that "it seems that even a very delicately determined proportional limit for a metal has little if any definite significance as an

index of strength or weakness." Moreover, the Committee report showed that, while short-time tests for tensile strength, yield strength, elongation, and reduction of area, made according to the provisions of the code, are reasonably reproducible in different laboratories, those for proportional limit show too wide a scatter to be dependable. Even if the proportional-limit determination were reproducible, its results would be directly

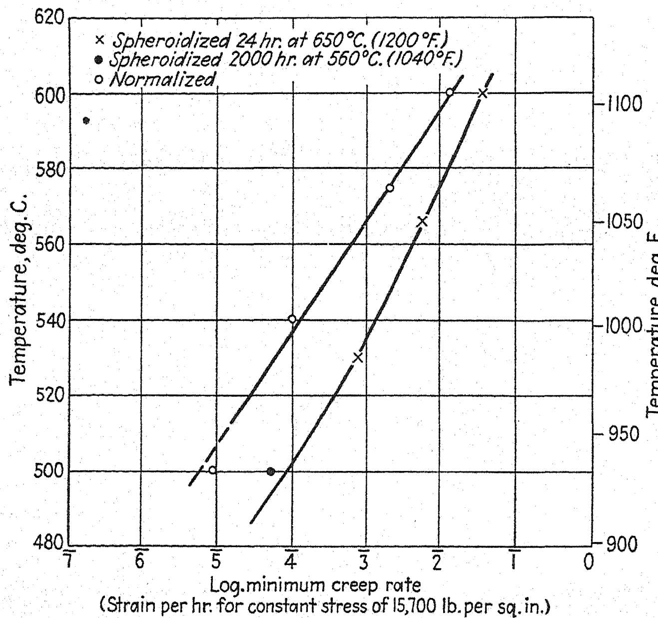

FIG. 192.—Creep rates for a 0.4 per cent carbon steel, as normalized and as treated to produce spheroidized cementite. (*Bailey and Roberts.*[507])

applicable only to a certain speed of loading which would be of no real engineering importance.

186. Deterioration of Creep Properties in Service.—Fortunately, the chief factor in the deterioration of carbon steels in high-temperature service appears to be that of oxidation of a type that can be observed, rather than that of a more subtle intercrystalline deterioration which affects some of the alloy steels used in high-temperature service. Another factor affecting the behavior of a carbon steel in service is that of spheroidization of cementite, the mechanism of which is better understood than is that of intercrystalline deterioration of alloy steels. Bailey

and Roberts[507] investigated the effect of spheroidization on creep in a 0.40 per cent carbon steel with the results shown in Fig. 192. These investigators also reported a marked drop in resistance in carbon and low-alloy steels after 2000 hr. at 560°C. (1040°F.). This was explained as a type of temper brittleness, which appeared even in steels which were not subject to temper brittleness on being maintained at temperature for normal tempering times.

C. CREEP DATA ON CAST FERROUS ALLOYS

As indicated in the previous chapter (page 482), differences in high-temperature properties between wrought and cast steels can be traced chiefly to the marked difference in grain size; much of the material in the following discussion deals, therefore, with the effect of this variable. As is the case in determining short-time mechanical properties, creep data of the greatest significance for a comparison of cast with wrought material are obtained by means of split heats.

187. Creep Characteristics of Cast versus Wrought Steels.—The high-temperature properties of cast and wrought carbon steels were determined by Spring.[415] Results of short-time tests are given in the previous chapter (see Figs. 175 and 176, pages 480 and 481). Spring also made creep tests. His values, given in Table 91 are the loads, in lb. per sq. in., to produce the specified elongations in 10,000 hr. at the various temperatures. The tests for the 0.20 per cent carbon steel were made on a split heat, part of which was tested as cast and annealed, the other part was forged and annealed. The elongation values for loads of 10,000 and 20,000 lb. per sq. in. at 425°C. (800°F.) and for loads of 3000 and 10,000 lb. per sq. in. at 540°C. (1000°F.), versus time in days for this steel are plotted in Fig. 193. Comparison of the data shows that the forged material had a slightly superior creep resistance at 290 and 425°C. (550 and 800°F.), but at 540°C. (1000°F.) the cast material had a decidedly higher creep resistance.

The creep data for the 0.35 and 0.40 per cent carbon-steel heats do not show such clearly defined trends. The tests, in this instance, were not made on split heats, and the cast and wrought steels were of slightly different composition.

TABLE 91.—CREEP DATA ON CAST AND WROUGHT CARBON STEELS*

Testing temperature		Stress (lb. per sq. in.) to produce an elongation of					
		0.1 per cent		1.0 per cent		10 per cent	
°C.	°F.	Cast	Wrought	Cast	Wrought	Cast	Wrought
0.20 per cent carbon							
290	550	28,000	32,000	37,000	38,000	46,000	45,000
425	800	8,000	9,000	15,000	16,000	27,000	23,000
540	1000	......	....	5,100	2,600	9,500	6,500
0.33 (cast) and 0.35 (rolled) per cent carbon							
290	550	......	22,000	....	44,000		
400	750	16,000	....	20,000	......	35,000	
425	800	10,000	8,000	17,000	23,000	23,000	
540	1000	1,800	1,300	3,500	4,300	12,000	10,000
650	1200	......	....	>1,000	1,000	2,200	1,000
0.40 (cast) and 0.41 (forged) per cent carbon							
290	550	15,000	....	22,000			
400	750	11,000	....	18,000	......	37,000	
425	800	8,000	7,000	17,000	25,000	23,000	35,000
540	1000	2,000	>1,000	3,000	2,500	4,800	8,000

* Spring.(415)

Cast specimens were annealed, and hot-worked specimens were normalized before testing. At creep rates of 0.1 per cent the cast material seems slightly superior; at higher rates the forged material, for most of the tests, shows a higher creep resistance, with the exception that the cast material is usually better at a temperature of 540°C. (1000°F.) or above.

The creep resistance of a 0.30 per cent carbon cast steel and a 0.40 per cent carbon forged steel was determined by Bailey and Roberts.(507) Results are as follows:

Specimen	Stress for a Creep Rate of 1 Per Cent in 10,000 Hr. at 500°C. (930°F.)
Cast	11,200 lb. per sq. in.
Forged	17,300 lb. per sq. in.

On account of the difference in carbon content and the fact that a split heat was not used, these results cannot be taken as conclusive in determining the relative creep resistance of truly comparable cast and wrought materials.

That cast steels are not necessarily superior in creep resistance at all elevated temperatures to wrought steels of the same com-

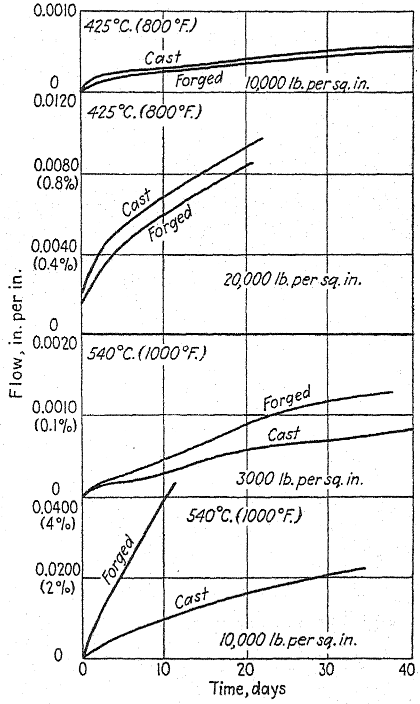

FIG. 193.—Creep characteristics of cast and annealed and of forged and normalized specimens from a split heat containing 0.20 per cent carbon. (*Spring.*[415])

position is shown by tests of split heats of austenitic (18 per cent chromium, 8 per cent nickel) steel[682] where rolled material showed, on the whole, better results than cast. As far as ordinary carbon steels in the usable temperature range are concerned the general opinion seems to be that cast steels have advantages, but the question needs further examination and clarification.

188. Effect of Grain Size on Creep Characteristics.—As noted in the previous chapter, causes for the differences in the high-

temperature properties between cast and wrought materials can usually be found in differences in grain size. Wilson[788] cited creep tests on carbon-molybdenum steels which indicate that for *high* creep rates at low temperatures coarse-grained steels are stronger, but for *low* creep rates at low temperatures—in this case below about 480°C. (900°F.)—the finer grained steels show higher creep resistance. At more elevated temperatures—in this case 540°C. (1000°F.)—the coarse-grained ones show slightly better creep resistance, while at very high temperature—650°C. (1200°F.)—the coarse-grained are materially better, the difference being accentuated at very low creep rates.

White and Clark[784] examined two molybdenum alloy steels, each in coarse- and fine-grained condition, and found that the temperature had to be somewhere between 425 and 540°C. (800 and 1000°F.) before the creep resistance of the fine-grained steel was surpassed by that of the coarse-grained one and that the difference between the two is often not sufficiently large to be of commercial importance.

White, Clark, and Wilson[785,786] in discussing the same data commented:

> The writers believe, however, the coarse-grained steels are not enough stronger than the fine-grained steels in the important range of industrial temperatures to urge a preference for coarse-grained steels. In some respects the fine-grained steels have advantages over the coarse-grained steels. They do not coarsen so rapidly at forging temperatures and are invariably tougher and more ductile at atmospheric temperatures.

Herzig, in discussion of White and Clark's paper,[784] suggested that the equicohesive temperature may be different in coarse-grained and fine-grained steels of the same composition.

The work of Enders,[694] also discussed by Körber,[728] on the problem of grain size, is unfortunately marred by the use of the unreliable "accelerated" method of Pomp and Enders, the lack of trustworthiness of which has been previously commented upon. If any value at all can be placed on Enders' estimate, from the rate between the 5th and 10th hour, of what the actual load might be to produce a final deformation rate of 1 per cent in 10,000 hr., his work would lead to the conclusion that a 0.37 per cent carbon, 0.31 per cent silicon, 0.67 per cent manganese rolled steel, studied at 500°C. (930°F.), increases its load-carrying

ability progressively from (*a*) 6000 lb. per sq. in., to (*b*) 9800, (*c*) 13,500, and (*d*) 15,000 lb. per sq. in. as the grain size is increased by overheating from the ordinary annealing temperature, used for (*a*), to annealing temperatures of (*b*) 1050°C. (1920°F.), (*c*) 1200°C. (2190°F.), and (*d*) 1300°C. (2370°F.). However, the two highest annealing temperatures produced not only a coarse but a Widmanstätten structure, with elongation values in the tensile test of only about 13 per cent. The impact values are not given but would doubtless be so low as to make such steel useless for many purposes.

In a lower carbon rolled steel of 0.21 per cent carbon, 0.17 per cent silicon, and 0.62 per cent manganese Widmanstätten structure was produced only by annealing at 1300°C. (2370°F.); the elongation values in the tensile test remained at about 27 per cent for all the other three treatments.

The estimated load-carrying ability at 500°C. (930°F.) for the material annealed at ordinary temperature was 12,400 lb. per sq. in., and for material annealed at all the other temperatures it was the same, 14,200 lb. per sq. in., although the grain size was decidedly increased at the 1200°C. (2190°F.) anneal over that at 1050°C. (1920°F.). Thus the 0.20 per cent carbon steel showed (by the test used) no such sensitivity to increase in grain size as was found in the 0.37 per cent carbon steel.

Enders made no direct comparison between cast and wrought steels of wholly analogous composition but did present "accelerated" test data on two cast steels of the following composition both being high in manganese for steel castings:

Element, per cent				
C	Si	Mn	P	S
0.20	0.29	1.05	0.052	0.058
0.17	0.37	0.83	0.08	0.015

In the green, as-cast state they had respectively only 4 and 8 per cent room-temperature elongation. The estimated loads for a final creep rate of 1 per cent in 10,000 hr. were respectively 16,400 and 15,000 lb. per sq. in. at 500°C. (930°F.). On annealing, the room-temperature elongation values rose only to 7 and

16 per cent respectively, and the creep loads were estimated at 11,700 and 12,000 lb. per sq. in.

Hence, of these cast steels, normally annealed, and the 0.21 per cent carbon rolled steel, also normally annealed, the wrought steel had a very slight advantage at 500°C. (930°F.) on the basis of the Pomp and Enders method of evaluation, although the micrographs show the cast steels to have much larger grain size than the wrought steel.

Since the method of evaluation is of doubtful validity, since the cast steels were not of acceptable ductility to represent ordinary cast steels, and since the effect of increase in grain size was so different in the two rolled steels, the chief value of Enders' work is in making it evident that too sweeping conclusions should not yet be drawn on the effect of grain size or the relative creep resistance of cast versus wrought steels.

It often occurs that oxidation makes a given type of steel commercially unusable before such a high temperature has been reached that a coarse-grained steel would be preferable. In judging the relative merits of wrought and cast steels, on the basis that cast steel usually has the coarser grain, it should not be forgotten that castings may be difficult to gate and feed so that they are free from incipient pipes or shrinkage cavities, blowholes, and similar defects, while wrought articles may normally be made quite free from analogous defects. There also seems to be a far greater tendency to "tension brittleness" in coarse-grained cast materials than in fine-grained wrought materials, the former sometimes showing very little creep for hundreds of hours and then failing suddenly by cracking without appreciable deformation. This behavior usually does not appear in all test specimens of a given lot and is therefore not characteristic of the material itself in its best condition but rather of an individual occasional test bar, or an occasional casting in service, presumably one that was not properly fed or in which some of the coarse crystals were poorly oriented to withstand the applied stress and so separated locally to form cracks. The chances for such behavior seem far smaller in a fine-grained material.

The broad statement that coarse-grained material is better for high-temperature service seems far from justified, even though it may be true for temperatures well above the softening tem-

TABLE 92.—CREEP DATA ON GRAY AND HIGH-TEST IRONS*

Iron	Rates of deformation, in. per in. per hr., in creep tests at 370°C. (700°F.) and a stress of 10,500 lb. per sq. in.				Impact at 370°C. (700°F.), ft-lb.†	Set under proof load at 425°C. (800°F.), in. per in.‡	Tensile strength at room temperature, lb. per sq. in.			Brinell hardness measured at room temperature	
	0 to 500 hr.	500 to 900 hr.	1100 to 1600 hr.	1600 to 2000 hr.			Before creep testing		After creep testing	Original	After 2015 hr. at 370°C. (700°F.)
							Maximum	Minimum			
Gray	0.00000121	0.00000059	0.00000034	0.0000003§	7.1	0.0051	33,900	31,500	29,750	197	156
High-test	0.00000053	0.00000053	0.00000034	Practically zero	7.7	0.0002	48,300	45,000	48,250	237	217

* Allen.(428)

† Large Charpy specimen 4.06 in. long, 0.80 × 0.80-in. section, span 3.20 in., V-notch 0.16 in. deep.

‡ Proof load of 25,000 lb. per sq. in. applied 10 min., load released, and permanent set measured at room temperature.

§ 1900 to 2000 hr., practically zero.

perature. Such statements require proof in respect to the exact material to which reference is made and should be related to the particular temperature at which service is contemplated.

189. Creep Tests of Cast Iron.—Bolton* found no definite evidence of creep at 315°C. (600°F.) with a load of 10,000 lb. per sq. in., within the precision of his test method, for a cast iron of 2.72 per cent total carbon, 0.58 per cent combined carbon, 2.32 per cent silicon, 0.24 per cent phosphorus, 0.052 per cent sulphur, and a tensile strength (0.75-in. diameter bar) of about 48,500 lb. per sq. in.

Creep tests at 370°C. (700°F.) with a load of 10,500 lb. per sq. in. were reported by Allen(428) in a discussion of Bolton and Bornstein's(437) article. The composition, room-temperature properties, and short-time tensile properties, of the gray iron and high-test iron are given in Table 89 and in Fig. 181 in the previous chapter. The creep rate, in in. per hr., and other pertinent data are given in Table 92. Curves showing rate of flow for these two irons and for two molybdenum cast irons are plotted in Fig. 75, page 189, of "The Alloys of Iron and Molybdenum."

D. AUTHOR'S SUMMARY

1. The use, in certain industries, of increasingly higher temperatures and increasingly heavier loads has made it very necessary that engineers be supplied with reliable data on the resistance of ferrous materials to slow deformation (creep) or to actual fracture when such materials are subjected to these high temperatures and loads for long periods of time. It is even more difficult to secure such data than to determine the usual mechanical properties at elevated temperatures, as the time factor—the rate of application of the load—becomes of major importance.

2. At room temperature and under low loads steel is essentially elastic, it undergoes plastic deformation only when subjected to loads above the elastic limit; plastic deformation becomes marked, however, only at stresses above the yield point. (Cast iron, on the other hand, exhibits plastic deformation at room temperature.) For most metals and alloys there is some temperature below which the behavior under low loads is essen-

* See discussion of Bolton and Bornstein's(437) article, p. 464 of Symposium.

tially elastic and above which it is essentially plastic. This dividing line is the softening temperature—called the equicohesive temperature—at which hardening from prior cold deformation is destroyed. As this softening is the resultant of both time and temperature, the temperature cannot be fixed accurately.

3. When steel is stressed at temperatures below the softening temperature, strain hardening occurs with the result that the rate of creep decreases despite some decrease in cross-section. Above the softening temperature, where annealing destroys the hardening from cold working as fast as it occurs, the decrease in cross-section is not compensated for by strain hardening, hence the rate of creep may finally increase.

4. Under conditions where strain hardening and a decrease in the rate of creep may occur there may be some justification for the use of the term "creep limit," to define an allowable degree of deformation. Above the strain-hardening range, however, it is doubtful whether there is any load that will not cause some minute deformation. If this is the case, and it is becoming increasingly apparent that some materials stretch continuously at these temperatures, there can be no creep limit; the important factor is whether the *rate* of creep is small enough for the purpose in hand over the period of service expected.

5. Because the above factors were not taken into consideration, and because of experimental difficulties, most of the older data on creep are of little value. Recently, a coordinated effort has been made to standardize creep testing. One result is that creep is being expressed not as creep limit, where all of the variables must be defined before the term has any meaning, but as the stress for a given "rate of creep," usually in terms such as 1 per cent in 10,000 hr. or 0.1 per cent in 10,000 hr. As the initial rate of creep is usually greater than the creep which occurs later, it is necessary to distinguish between total creep, which includes this initial deformation, and the final rate of creep, which excludes it. Recent data usually represent the final rate of creep.

6. Much time and effort have been expended in an attempt to devise an accelerated test whereby the creep rate after thousands of hours can be predicted from tests made over a few hours or, at most, a few days. So far these attempts have not been successful. Some of them apparently have value as qualitative

tests, or as acceptance tests to discover if different lots of steel of the same composition and history act alike in creep. Even for such purposes, however, accelerated tests should be used with caution.

7. There are few reliable data on the creep of carbon steels; most published results on creep were obtained on heat-resisting alloy steels. Most of the results available on carbon steels indicate that, in general, the creep resistance, at least in steels containing 0.6 per cent carbon or less, increases with the carbon content for temperatures up to 425°C. (800°F.). If the available data on the creep of carbon steels are plotted, considerable scatter becomes apparent. A notable part of this scatter is probably due to variations in the testing methods; a considerable part, however, is most likely the result of inherent differences in the steels themselves, differences which cannot be traced to composition as usually determined.

8. The creep properties of carbon steels as a class appear to be predictable only within limits which are far too wide for successful use by the engineer or designer who wishes to use carbon steels for service requiring high temperatures and sizable loads.

9. There is apparently no definite relation between proportional limit at high temperature, even if determined by the most precise methods, and the creep resistance of a carbon steel. The proportional-limit versus temperature curve is not parallel to the curve showing the stress required for a given amount of deformation in a given time. Moreover, the values for elevated-temperature proportional limit are rarely reproducible by different investigators, and, because they are obtained only by means of a definite speed of loading, are of no real engineering importance.

10. Although there are data, obtained on split heats, which indicate that, for low- and medium-carbon contents, cast steels have better creep resistance than wrought material at temperatures of about 540°C. (1000°F.), further confirmation is needed before this conclusion can be accepted as definitely established. Furthermore, the statement sometimes made that coarse-grained steels have better creep resistance than fine-grained material of the same composition is far too sweeping, even though it may be true for certain temperatures. Such a statement requires more evidence of its validity than is given by the data available now.

11. The status and value of creep testing for ferrous materials may be summed up with the statement that: (*a*) at present, no valid generalizations can be made on the relation between creep resistance and other readily determinable properties; (*b*) to be useful in design for high-temperature service, special creep tests must be made on the specific materials and at the specific temperatures under consideration; (*c*) large, and sometimes uncertain, extrapolations must be made to predict the behavior of a material over many years of service; (*d*) the question must be answered whether the material undergoes slow structural changes during long exposure at elevated temperatures; (*e*) the effect of oxidation and the corrosive effect of the atmosphere surrounding the material in service and at the service temperatures must be known; (*f*) the actual temperatures of service must be known with much more accuracy than at present; (*g*) the tendency toward creep should be separated from the other variables influencing the life of a material at elevated temperature; and (*h*) laboratories should continue their search for the principles underlying creep.

CHAPTER XIV

THE CORROSION OF COMMERCIAL IRON-CARBON ALLOYS

Types of Corrosion—Measuring Corrosion Resistance—Influence of Composition on Corrosion—Oxidation at Elevated Temperatures—Author's Summary

In the presence of air and moisture, iron and steel rust. It is only in those extremely rare cases of industrial use where moisture is not present in the atmosphere, or oxygen in the water, that ordinary steels and irons of commerce are free from corrosive attack. Either the use of some type of protective coating or treatment of the corrosive environment so as to make it non-corrosive is ordinarily required to overcome the inherent tendency of commercial iron-carbon alloys to corrode.

Many estimates have been made of the huge annual monetary loss from rusting of ferrous materials—a recent estimate by Mears[638] placed the loss in the United States alone in excess of a quarter billion dollars annually. Certainly, the loss is stupendous, and in depression years more iron may be reverting to its oxide through attack of the elements than is being extracted from ore.[776] The economic importance of this loss has led to much careful study of the corrosion problem; Rabald's[481] recent book contains 2800 references to the literature on the subject of iron and its corrosion.

It is not possible now, nor is it likely to be possible in the future, to fix a quantitative evaluation of the corrosion resistance of commercial iron-carbon alloys for use as a base-line in the determination of the resistance of other alloys of iron. Corrosion is a function of both the material and the environment, and no ferrous material is resistant to all environments. Hence, all that needs be done in this series of monographs is to select from the available information those data which indicate the relative behavior of alloyed and unalloyed ferrous materials or the relative behavior of different alloyed materials. The place

for this is in the monographs on the individual alloying elements. For this reason, only a brief and general treatment of the corrosion problem is presented here, especially since the corrosion of iron itself, the mechanism of the corrosion process, the topic of passivity, of solubility in chemical reagents, and the effect of impurities and added elements are discussed in the monograph "The Metal—Iron."[794] Corrosion is also treated at length in several other books, such as those by Friend,[19] Pollitt,[106] Cushman and Gardner,[16] Calcott, Whetzel, and Whittaker,[98] Turner and Budgen,[202] Kröhnke, Maas, and Beck,[314] De Lattré,[215] Pålmaer,[478] Rabald,[481] and Abegg,[675] while some comprehensive treatises of prime importance are those of Evans,[176] Speller,[200] and McKay and Worthington.[822]

A. TYPES OF CORROSION

Of the two types of corrosion, general and localized, the latter is the worse; it is bad enough to have an object rust away uniformly, but it is even worse to have it rendered useless by a localized attack that may not be readily detected. Iron and steel, save in rare instances which depend on the environment rather than on the material, are not subject to intergranular penetration and brittleness, but they may exhibit localized corrosion or pitting, which again may ordinarily be attributed to the environment rather than to the nature of the material. In the following discussion "iron" is used as a generic term relating to both iron and steel unless other usage is evident from the context.

190. General Attack.—A "tight" oxide coating, as long as it is unbroken, offers considerable protection, as shown by "Russia iron" in stove pipes, articles given the Bower-Barff coating, and blued gun barrels. Under most conditions of use, the natural rust coating on unalloyed irons or steels is not tight or impervious, the corroding mediums penetrate the rust coating and continue to form rust which may in time peel off in flakes.

Factors in the environment which accelerate corrosion are sulphur gases from fuel and the ever-present traces of chloride ions in solutions. These ions penetrate rust and oxide coatings with great facility. By immersion in chromate or dichromate solutions iron or steel becomes "passive," probably because it

becomes coated with a very thin film of oxide, but chloride ions break through this film with ease. Iron becomes passive when immersed in concentrated nitric acid; the surface film so produced is, however, easily broken down so that this treatment does not offer permanent protection against other reagents.

It is well known that acid waters corrode iron rapidly while moderately alkaline solutions are relatively inert. Sodium carbonate solutions are used as cooling mediums in machining steel because of their relative freedom from corrosive action. This can be generalized, in terms of hydrogen-ion concentration (pH) of the waters, into the statement that acid waters of pH below 4.3 corrode or, more accurately, dissolve the metal rapidly with evolution of hydrogen; the more nearly neutral waters of pH 4.3 to 10 cause rusting at a rate dependent on oxygen content, rate of flow, etc., while the more alkaline waters of pH above 10 have but slight action. Other substances in natural waters have an effect; magnesium chloride is especially harmful, while salts which deposit in a coherent coating over the iron may protect it. Organic colloids may exert a protective action. The behavior of any natural water, rain, or dew in a contaminated atmosphere is therefore a matter dependent on its composition.

At higher temperatures, strongly alkaline solutions attack iron. According to Parr and Straub,[193] sodium carbonate in crevices of boilers decomposes under the conditions of service to give sufficient caustic alkalinity to exert strong corrosive action with evolution of hydrogen and the production of "caustic embrittlement" by grain-boundary attack or intercrystalline penetration. It is claimed that the presence of a suitable proportion of sodium sulphate prevents this. Jones[71] showed that other reagents in addition to alkalies may cause intercrystalline attack. The whole question of caustic embrittlement has been recently reviewed by Partridge and Schroeder.[746]

Except in such special cases as that of caustic embrittlement, the rusting of iron does not ordinarily cause intergranular penetration and brittleness; the effect of corrosion is, therefore, chiefly that of a diminution of cross-section or the formation of pits. The condition of a rusted piece of iron or steel can, therefore, usually be told simply by scraping off the corrosion products and inspecting the piece; this is not possible in the case of some non-ferrous alloys which may look fairly sound on the surface

but be weak and brittle below. It is important, in the study of corrosion, to differentiate between general corrosion and localized attack by pitting. The latter is the more insidious.

191. Electrochemical Potential and Corrosion.—There is one inherent and measurable property of iron which affects its usefulness under commercial corrosive conditions and accounts for its protection by contact with a few other metals, which are sacrificially corroded instead of the iron, and for its accelerated corrosion when in contact with some other metals. This is the actual electrochemical potential of iron in contact with the solution, this potential being a measure of the tendency of iron to corrode in that solution, though not of the rate at which the corrosive process will continue. The actual potential depends upon the effective concentration of ferrous iron in the solution; the potential computed on the basis that this effective concentration is unity (that is 1 g. atom Fe^{++} per 1000 g. water) is the so-called "standard potential." The order of the metals in terms of the standard electrochemical potential is given in Table 93. This represents the order to be expected under most conditions, though there may be a reversal of position between elements near together in the series, by reason of marked disparity between the concentration of the respective ions or of overvoltage phenomena. For instance, cadmium would be anodic to iron in a solution containing a very high concentration of Fe^{++} and a very low concentration of Cd^{++}; tin may become less noble than iron in a solution containing little or no oxygen, as in canned food products.

If iron is coated with a metal below it in the series—*i.e.*, with a metal more noble than itself—such as nickel, lead, or copper, and this coating is porous, the iron at the bare spots should, and generally does, corrode more rapidly in the presence of an electrolyte than does uncoated iron.

The expectation that porosity will lead to rapid electrolytic attack owing to the local galvanic couple is, fortunately, often not realized in actual service, provided that conditions are favorable for the choking of the pores with rather impervious plugs of corrosion products. As an example, the atmospheric-exposure tests of the American Society for Testing Materials[719] have shown that a rapid attack occurs upon porous lead-coated steel in sea-coast, or even in rural, atmospheres, but in a highly

sulphurous industrial atmosphere the pin holes become plugged, apparently with sulphide, and the coating affords good protection.

Unless choking of the pores by corrosion products occurs in such fashion that contact of the anodic metal with the electrolyte is avoided, such a combination of a metal pair and an electrolyte accelerates corrosion. On the contrary, a coating with a metal above iron in the scale of Table 93 need not be wholly impervious to afford protection, since the electrolytic couple set up, while it

TABLE 93.—THE ELECTROCHEMICAL SERIES* OF THE METALS WHEN THE EFFECTIVE CONCENTRATION OF THE METAL ION IS UNITY†

Alkali metals—Cs, Rb, K, Na, Li
Alkaline-earth metals—Ba, Sr, Ca
Magnesium
Aluminum
Manganese
Zinc
Chromium‡
Cadmium
Iron
Cobalt
Nickel
Tin
Lead
Hydrogen
Copper
Arsenic
Bismuth
Antimony
Mercury
Silver
Palladium
Platinum
Gold

* In this series a metal placed above another is generally reactive (anodic) to the metal below it, in the presence of most ordinary electrolytes.

† "Handbook of Chemistry and Physics."(802)

‡ The position of a metal like chromium which can occur in the active or passive state will depend upon whether the measurements are made upon the true metal, or active, surface or upon the passive, film-coated, surface.

accelerates the attack on the protective coating, makes the iron cathodic and protects it, as on the cut end of a barb in galvanized barb wire. The size of the protected area varies with the electrolyte.

That these electrolytic phenomena explain many of the phenomena of corrosion under service conditions is undisputed.

The failure of steel parts in sea water in contact with Monel metal is a good example of the effect of electrolytic action.

Watts,[820] however, has kept specimens of Monel cut from one of the original plates of the yacht *Sea Call*, soldered to iron and to steel, under de-aerated sea water without sign of corrosion. From this and many other experiments with single metals and voltaic couples made up of various different metals and in various electrolytes (among them the iron-copper couple in ammonium chloride solution) he concluded that when certain electrolytes are *completely* freed from dissolved oxygen there is no electrolytic attack. This in no way alters the observed fact that under normal conditions of service corrosion is often accelerated by contact with another metal, but it may help to explain the many cases where such corrosion is expected but is actually not serious. The ability or lack of ability of the environment to supply oxygen evidently plays a considerable part. Schikorr,[339,340] however, has made experiments which indicate that steel is attacked by water with the evolution of hydrogen, even in the absence of oxygen.

Metal deformed by cold work, or metal in a stressed condition, is anodic with respect to undeformed, strain-free metal and therefore corrodes more readily. The electrolytic theory of corrosion applies to such cases and explains why the attack is more rapid on cold-worked rivet heads than on the plates they join. On the other hand, uniformly cold-worked metal by itself, *i.e.*, not in contact with another metal, does not necessarily corrode faster than the corresponding annealed material.

Johnston[724] has given a clear discussion of the limitations under which electrode-potential data may be applied to the prediction of the behavior of metal couples. The electrochemical properties of iron, including its position in the electrochemical series, its electrode potential, its overvoltage and passivity, have been discussed in detail in "The Metal—Iron."[794]

192. Localized Attack.—Pitting is a more dangerous type of attack on iron and steel than that of slow and uniform wasting away of the surface. Pitting is not usually a result of the chemical composition of the steel. In some cases it may be due to porosity or other physical imperfections in the metal itself, but it is more commonly due to external causes, such as the breaking of a protective film, the presence of mill scale, or contact of

two pieces in such fashion as to form a capillary space between them, leading to localized "crevice corrosion" from an oxygen-concentration cell. Pitting was produced by many soils in the National Bureau of Standards'(267,317,468,632,734) soil-corrosion work, but it was a function of the soil conditions rather than of the metal composition.

Localized attack may result from corrosion attributable to oxygen-concentration cells such as occur in "water-line" corrosion, "crevice" corrosion, and corrosion due to immersion at various depths in a solution where oxygen diffusion is hindered and the electrolyte in contact with the iron has a different oxygen content at different places. This is thoroughly discussed by many modern writers on corrosion and needs mention here only to emphasize that this prevalent cause for local corrosion or pitting is a part of the environment and entirely external to the metal.

An analogous cause of corrosion is the presence of discontinuous patches of mill scale on the surface of iron or steel, which localizes corrosion at the edges of the scale. The behavior of inclusions embedded in the metal and bared at the surface is less clear. Manganese sulphide inclusions are generally considered to accelerate corrosion to a small degree so that high-sulphur steels such as ordinary screw stock are classed as more corrodible than low-sulphur steels.

Other types of local attack may be connected with erosion as well as corrosion, and differences in oxygen concentration of the surrounding fluid may also enter. Perforation in water pipes may be connected with the "impingement effect," and corrosion of propellers is sometimes related to cavitation effects. The "hot-wall" effect described by Benedicks(149) may also enter in some cases of local attack.

193. Corrosion Fatigue.—Corrosion is especially dangerous when it occurs simultaneously with repeated stress.(458,472,504,523) "Corrosion fatigue" affects corrodible alloys to such an extent that under corrosive conditions they seem to have no true endurance limit. Heat-treated steel of 120,000 lb. per sq. in. endurance limit in air may show only about 15,000 lb. per sq. in. when tested in a stream of water. Strength due to alloying or heat treatment is not maintained unless the alloying element also increases the resistance to corrosion, and all the corrodible

ferrous alloys tend to come down to the same mediocre level. This is discussed in Chapter XI, page 419.

B. MEASURING CORROSION RESISTANCE

Were it possible to measure, in exact and absolute units, the real corrosion resistance of a metal or alloy in a manner analogous to that by which density, magnetic properties, etc., may be measured, there would be no data more useful than those on corrosion resistance. But this cannot be done. Corrosion resistance is inevitably a comparative and relative thing, definable only in terms of factors relating far more to the environment in which the metal is used than to the factors dependent on the metal itself. It cannot be stated that a 16-gage sheet of normalized 0.10 per cent carbon steel *will* fail on atmospheric exposure in X years, or that a wrought-iron or steel water pipe *will* fail in Y years. It can be stated only that certain materials, under the exact environment then existing, *did* fail in a given period. It is not certain that this exact environment will be met in any future service of an analogous material. In cases of rapid chemical solution, one might be able to fix with some degree of exactness how long ordinary iron or steel would last, but it would be a useless calculation, since these materials would not be serviceable under conditions where their life is so short that it can be accurately predicted. Actual use of the materials is confined to cases where experience has shown that in spite of slow attack an economically profitable life may be expected.

194. Accelerated Tests.—Myriads of "accelerated corrosion tests" have been made, under more or less carefully controlled conditions, with results expressed in definite units, but these units are not transferable into years of service under other conditions. Such tests may give useful comparisons from which, if proper experience and engineering judgment are applied, qualitative expectations of relative service may be gathered provided the test conditions have only accentuated, but in no wise altered, the factors governing the corrosion in actual service.

A more reliable approximation to expected life under defined conditions may be obtained from non-accelerated long-time exposure tests in which a group of materials is exposed in the same places for the same periods, and careful periodic inspections are made during the period of years required for natural condi-

tions to result in failure. Such tests, notably those conducted by the American Society for Testing Materials, by the National Bureau of Standards, and by the British Corrosion Committee,[459] are adding much to our knowledge of corrosion resistance. All such tests emphasize the fact that, as far as ordinary unalloyed iron and carbon steel are concerned, variations in environment are of greater import than variations in the ferrous material.

Whether or not wrought iron, ingot iron, low-carbon steel, and copper-bearing steel and its modifications differ materially in corrosion resistance under commercial conditions has been a much discussed subject, with many of the discussions inspired by commercial considerations rather than by the truly scientific findings of the respective producers. The now thoroughly discredited "test method" of determining rate of solution in acid when service in air or in fresh or salt water was contemplated has beclouded the issue until the American Society for Testing Materials[246] had to issue the following statement:

Notwithstanding the fact that this Committee, in its annual report of 1909, pointed out as clearly as it could that the tentative suggestions made by it in 1907 as to the conditions for carrying out the so-called acid corrosion test were not to be considered a recommendation of the test, and that the results of such test are unreliable as truly measuring the tendency to natural corrosion, the name of the American Society for Testing Materials continues to be used as having recommended the acid corrosion test, and by inference as having endorsed the same. For this reason the Committee desires at this time to again disclaim any recommendation or endorsement of the acid test as a measure of natural corrosion, and to point out that any use of the name or authority of the American Society for Testing Materials in this connection is unwarranted.

Evans*[176] discussed the various accelerated tests, including salt spray, that have been proposed for information on behavior under certain marine conditions, and immersion tests under varying conditions of aeration to give information for particular types of service. He disposed of the so-called acid-immersion test by stating that it would be perfectly sensible *if* the order of resistance to acid were that of resistance to water and oxy-

* Pages 189 to 191 of his book.

gen. It is well known that this is far from being the case. He remarked:*

Before designing a technical test to estimate the value of a material for an engineering purpose, it is necessary first to study very closely the practical conditions of service, and then to try to reproduce them as far as possible in a laboratory test. Only if this is done will the test give reliable information. As stated by Speller: "There is no all-round test. . . . Accelerated tests should have a scientific and practical connection with service tests."

Evans believes that qualitative information on corrosion of metals by liquids with which they are to come in contact may be secured by the partial-immersion method (which brings out water-line effects) if the experimenter intelligently observes and correlates the phenomena that occur.

The Navy Department uses a salt-spray acceptance test for comparison of deliveries (of stainless steel in particular) with the known behavior of acceptable material, in that test, but stated:(350) "No accelerated corrosion test will give definite information as to satisfactory life in service, as it is impossible to foresee or duplicate service conditions."

The acid-solution test showed a relative corrosion resistance in acid for ingot iron and copper-bearing steel which, when misinterpreted as being applicable to corrosion resistance under other than acid conditions, far overrated the copper-bearing type. Although the overzealous advocates of copper steels convinced only the unthinking by the false argument of the acid-solution test, there is other, and real, evidence, especially in the comprehensive tests of the American Society for Testing Materials, that copper is definitely helpful in improving corrosion resistance, particularly under severe conditions of atmospheric exposure. This has been discussed in detail in the monograph on "The Alloys of Iron and Copper."(705)

195. Long-time Tests.—While so many environmental factors enter into corrosion that one can seldom list them all in any particular case, much less do so for all corrosive conditions, a few vital factors are generally recognized. The amount of wetting is important. Exposure in a dry atmosphere, exposure in alternately wet and dry conditions, as when the material is wet by

* Page 193 of his book.

rain or dew and then dries out, immersion without opportunity to dry out, and in total immersion whether the water is moving or stagnant, all present different conditions. The corrosion products, hydrated iron oxides, are rather gelatinous and may adhere or be washed off depending on conditions. If they are not washed off and can dry down to a hard rust, they may afford some protection against further access of the corroding mediums. The composition, physical properties, and adherence of the initial rust coating are of prime importance in what happens thereafter.

The pillar of Delhi is often cited as an outstanding example of corrosion resistance; it was erected about A.D. 300, but Hadfield[(156)] brought a piece of it to England where it rusted overnight when wetted. Its resistance is undoubtedly largely due to the purity and dryness of the air of its environment and perhaps to the nature of the initial corrosion products.* Steel articles left at Panama by the French from their work on the Canal have been found intact after many years but, when the old coating of corrosion products was removed, rusting went on at an extremely rapid rate.

Evans†[(176)] commented on relative freedom from rusting of iron and steel in pure mountain air, as in the Alps. The resistance of the ship plates of the *Leviathan* is ascribed by Rawdon[(334)] to the formation of initial "tight" corrosion products.

It is, therefore, apparent that the probable rate of corrosion or rusting in any environment can be estimated only from long-time tests made in the *same* environment, and that the attack depends not only on the surroundings but on the condition of the surface of the material at the beginning of the exposure. Nevertheless, some extrapolation is warranted, and it may be assumed with some safety that *relative* rates of continuance of corrosion after the initial period will not in general be materially changed by slight changes in the conditions of exposure.

196. Simulated Service Tests.—Atmospheric-exposure testing, like any other test method, has a technique of its own, though one not so thoroughly standardized as in many other cases. First, representative geographic locations must be chosen for the exposure tests, then the selection of specimens of thoroughly

* The corrosion of ancient iron is discussed in detail in "The Metal—Iron,"[(794)] pp. 300–302.

† Page 150 of his book.

known history, their cleaning or other preparation to produce an initial surface in a known, and reproducible, condition, the details of the method of supporting the specimens, the duration of exposure, the recording of the weather conditions, the atmospheric contamination by combustion gases and soot, the periods of dampness and dryness of the specimen, the temperature, etc., and particularly the number and size of specimens required to give dependable results according to statistical methods must all be considered.

The evaluation of the progress of corrosion by visual examination, by weight gains and losses (including choice of cleaning methods), by change in tensile strength, electric resistance, or other physical properties, by measurement of depth of pits, etc., brings in many important questions.

Similar complex problems arise in controlled-condition simulated-service tests in fresh and salt water. In any case all these matters have to be resolved with particular reference to the specific object of the test. The objects of simulated-service testing, therefore, are not served by merely putting out a few specimens and weighing them after a year or so. The rate of attack is often as important as the total attack in a given period so that many observations are required, especially in the initial period. Such testing, properly done, is by no means as inexpensive as it sounds. These matters have been discussed in detail by Evans,[176] Speller,[200] the Corrosion Committee of the British Iron and Steel Institute,[459] and in many of the reports of the various A.S.T.M. corrosion committees.

Interpolation of the A.S.T.M. exposure-test data on bare steel, zinc-coated steel, non-ferrous metals and alloys, and pairs of dissimilar metals, so as to evaluate not only the behavior of the metals but also the relative corrosiveness of the various locations at which the exposures were made, was attempted at a Symposium of the Society. Papers by Finkelday,[698] Hocker,[719] Hippensteel,[718] Passano,[748] and by Schramm and Taylerson[767] at this Symposium, as well as the summary by Mutchler[743] of comprehensive exposure tests on duralumin, allow a rough comparison of the corrosive activity of the atmosphere at the locations used. Broadly speaking, tropical marine atmospheres and highly polluted industrial atmospheres are highly corrosive, while a dry rural atmosphere is of course the least corrosive to

most metals, with moderately polluted industrial and non-tropical marine atmospheres in between. It should be emphasized that the order of corrosiveness of a given list of localities varies according to the type of material which is being corroded, that is, according to the chemical reaction and the nature of the corrosion product formed. A location such as that of New York City, where the atmosphere contains both SO_2—from the industrial use of fuel—and chloride—from the sea—is especially prone to show rapid corrosion. Inland industrial localities are more severe for both bare and coated steels than are non-industrial sea-coast localities, but this is not necessarily true for non-ferrous materials.

197. Requirements for Accelerated Corrosion-test Methods.—The experimenter with new alloys is justified in seeking preliminary information, prior to long exposure or to simulated-service tests, so that he may eliminate from his long-time tests those alloys which are of no promise for the purpose in hand. But, if his rapid pilot tests are to be reliable guides, he should make sure that the same chemical reactions are involved in the tests as in service, and that the physical state, the chemical composition, and the adherence and distribution of the corrosion products are the same as in service. In atmospheric corrosion there are normally idle times, for example, when moisture is not present and when corrosion does not advance. These periods might well be eliminated in an accelerated simulated-service test in order to save time, provided that they do not, by some alteration of the corrosion products, produce a tight, impervious coating resistant to further corrosion. If they do produce such a coating, idle periods sufficient to bring about these changes in the corrosion products should be included in the accelerated testing cycle.

When all the rapidly corroded test specimens of a series are indistinguishable, by appearance or by any other criterion, from similar specimens slowly corroded in service, the results of the rapid test would deserve respectful consideration. Unless they are indistinguishable, the results are likely to be misleading.

Bengough[592] stated that an acceptable accelerated corrosion test is one which reproduces the corrosion-versus-time curve of service except that the time scale is compressed.

C. INFLUENCE OF COMPOSITION ON CORROSION

There has been much discussion about the relative corrosion resistance of the common ferrous products, but it is apparent that they all rust so rapidly when exposed to the action of the weather and other corrosive mediums that it is difficult to say which of these materials is the worst. Some of the observations on the corrosion resistance of these materials which are decidedly pertinent are touched upon below.

198. Relative Corrosion Resistance of Common Ferrous Materials.—There is no serious claim that ordinary low-carbon steel is actually superior in corrosion resistance to wrought iron or ingot iron of very low carbon content. Its advocates merely claim that it has the same, or such similar, corrosion resistance that it would be advantageous from the engineering point of view to spend a given sum of money for a heavier sheet or pipe of the cheaper steel than for a thinner one of the more expensive material. Leaving aside the honestly differing opinions of prominent metallurgists connected with manufacturers of these rival materials, we find that those who have studied the corrosion problem from an impartial user's point of view are often in doubt which material is superior. In 1891 Howe(39) canvassed shipbuilders and railroad men for their opinion based on service records and found opinions quite equally balanced. Friend*(19) and Evans†(176) consider that there is little difference between wrought iron and steel. Hadfield,(86) a steel maker, believes that wrought iron is superior in sea water. Curran and Sanford(599) said of steel and wrought iron: "In our own experience, the two materials appear approximately equal, with perhaps a slight advantage, under some conditions, in favor of wrought iron."

The American Society for Testing Materials' atmospheric and fresh- and salt-water immersion tests(172) afford no clear evidence of superiority of one or the other of the ferrous materials under test, except for copper-bearing steel in the atmosphere, though there were great differences in the rate of attack and the total useful life in different locations and different waters.

The extensive soil-corrosion tests of the National Bureau of Standards(267,317,468,632,734) have shown no marked superiority for

* Page 286 of his book.

† Page 186 of his book.

any of the commonly used pipe materials—wrought iron, cast iron, and steel—in underground corrosion on the basis of the first 10 years' tests, though enormous differences were shown in the corrosiveness of different soils with all the materials. From the study it was concluded:

> On account of unavoidable variations in soils of the same type, in metals even of the same kind, and in methods of construction, exact rates of corrosion cannot be predicted, but approximate rates of corrosion can be given for specified metals and soil conditions.
>
> Differences in the rate of penetration of different pipe-line materials by soil action in the same soil are much smaller than differences in the rate of penetration of the same material in different soils. The type of soil rather than the variety of ferrous material is usually the controlling factor with respect to corrosion. In certain soils, however, one type of material may corrode much more rapidly than some other material and for this reason the soil to which it is to be exposed should be considered in selecting material for a pipe line.

This is again emphasized in the National Bureau of Standards' report by Denison and Hobbs,[686] who began their report with the statement: "One of the most important conclusions which has been drawn from the National Bureau of Standards' investigation of underground corrosion is that corrosion is determined by the nature of the soil rather than by the kind of ferrous material exposed to the soil." They further pointed out that the "corrosion pattern," *i.e.*, the shape and depth of the corroded areas, is similar for different ferrous materials in a given soil, probably being a function of the anodic and cathodic areas set up by the action of the soil, and that when, by the lapse of years, the soil becomes uniformly packed so that its contact with the specimen, the moisture at the contact surface, and the degree of aeration are all uniform over the surface of the specimen, the attack will become more uniform rather than localized, this leading to the idea of surrounding buried pipes with material of small particle size which will pack uniformly.

From this work it appears that if the soil is inhomogeneously packed, so that there are corroded and uncorroded areas, the attack upon the areas which do corrode is of the pitting type. In some measure it seems that a certain amount of corrosion will take place anyhow and it is highly desirable that this be distributed as a uniform surface attack rather than localized as pit-

ting. The British Corrosion Committee[459] also has extensive experiments in progress which should clarify the knowledge of the effect of copper in steel and the influence of the surface condition at the time of exposure.

Since wrought iron is credited with taking a thicker coating in the hot-galvanizing bath than steel, and since it is universally accepted that, if the zinc coatings are each of uniform quality, the thicker the coating, the longer its life,[585] wrought iron may have some advantage on that score.

Whether there is any difference in the rate of corrosion of an initially galvanized surface of iron and one of steel after the zinc coating has weathered through in patches is a matter of dispute. The American Society for Testing Materials has tests in progress designed to throw light on this question.

Although from its heterogeneity one might expect that ordinary cast iron would be peculiarly subject to corrosion, most authorities agree that it is one of the most lasting of ferrous engineering materials. This is at least partly ascribable to the greater thickness of cast iron as compared with competing ferrous materials, so that much greater loss can be borne before failure.

Friend[19] found in long-time atmospheric-exposure and sea-water tests at four widely separated localities that cast iron suffered less loss in weight than wrought iron, Swedish charcoal iron, open-hearth iron, or several carbon steels.

Speller*[200] ascribed the long life of cast iron in the atmosphere to a more adherent type of rust. Evans†[176] cited cases of remarkable life of cast iron in sea water but commented that, on account of "graphitic softening" (removal of iron, leaving graphite in place) which may occur in sea water with some varieties of cast iron, it is a material that "requires close watching." While it is generally believed that close-grained iron is superior in corrosion resistance to open-grained iron, there is little available information on the relative behavior of different types. It would be of decided interest to know under what environments modern high-test iron is better or worse with regard to corrosion resistance.

Schwartz[94] dealt but briefly with the corrosion resistance of malleable iron, citing the fact that service failures of malleable

* Page 266 of his book.

† Page 188 of his book.

iron by rusting are rare and commenting upon a relative service test of malleable iron and steel railway tie plates in which the malleable iron outlasted the steel. Wolf and Meisse[500] made some comparative exposure tests under the action of locomotive smoke in which malleable iron showed slight superiority over wrought iron, ingot iron, and basic open-hearth steel.

199. Influence of Elements Present in All Ferrous Products.—Since there is evidence that increased purity markedly increases the corrosion resistance of some metals, the question arises whether really pure iron, were it commercially obtainable, would be a corrosion-resistant material under actual conditions of use. The classical example is the hydrogen-reduced iron of Lambert and Thomson,[17,24] discussed by Bancroft,[113] which was stated not to rust in contact with purest distilled water and purest oxygen. It was not resistant, however, unless the oxide from which it was reduced was most carefully freed from all the ordinary impurities of commercial iron; also it was not resistant when traces of a more noble metal such as platinum were present or when the pure iron was strained. Evans[176] remarked that, while electrolytic iron stored without free access to the atmosphere may resist rust for years if the iron is deposited from a sulphate bath, iron made from a chloride bath, and hence containing traces of chloride, rusts with extreme rapidity, an observation that all who have used the latter material can verify.

There seems no hope of making iron so pure that it will really act as a noble metal for exposure under commercial conditions which involve contact with soot, traces of chlorides, oxygen-concentration cells, etc., and Evans considers that high purity need not be a necessary criterion of even moderate resistance. He stated: "On the whole the hope of obtaining corrosion-resisting substances lies in the addition of suitable constituents rather than in the removal of unsuitable ones." It seems probable, even if the purest iron were proved to be highly resistant to the initiation of rusting, that, once rusting had begun through the agency of some external cause, it would then continue to rust as rapidly as does commercial iron.

Cleaves and Thompson, in "The Metal—Iron,"[794] summed up the effect of composition by stating:* "On the whole, it

* Page 300.

appears that purity is not the determining factor in resistance to atmospheric corrosion." The conclusion is also reached* that "apparently, small amounts of carbon, 0.2 per cent or less, are of negligible importance to the corrodibility of iron. Larger amounts increase the tendency to corrode, the extent of the increase being affected by the state of combination of the carbon." Writers on corrosion generally agree with Evans†[176] that, while acid attack upon steel increases slightly with increase in carbon, increasing the carbon content from 0.05 to 1.46 per cent has little effect on the rate of rusting of steel by fresh or salt water containing oxygen. That different metallographic constituents in quenched steel, untempered or subjected to various tempering treatments, etch differently in acids is, of course, well recognized by metallographers. Speller‡[200] found that there was little difference in the rate of corrosion of steel in different conditions of heat treatment, of different carbon content or grain size, and cited several other investigators who came to similar conclusions. Cementite itself is less readily attacked than ferrite, but when the two are in contact, the cementite may accelerate the attack on the ferrite.

Manganese appears to have no clearly discernible effect, nor have normal amounts of phosphorus. Within usual limits, sulphur has little effect; but large amounts, as in screw stock, accelerate rusting. There are few data on oxygen content. Ingot iron, containing more oxygen than other commercial ferrous materials, is often classed among those which show relatively low susceptibility to corrosion; on this basis oxygen content can hardly be a vital factor in accelerating corrosion.

Stoughton, in reviewing an earlier draft of this chapter, commented that he does not accept this basis of argument, at least as it applies to steel. He stated that he has personally known of severe corrosion and pitting in highly "oxygenated" steel and believes that such steel will corrode more severely than steel free from oxygen or oxides and that this is generally considered to be a fact. Stoughton includes oxygen in blowholes or some complex inclusions, as well as that present as ferrous or manganous oxide, in the term oxygenated.

* Pages 326 and 327.

† Page 185 of his book.

‡ Pages 78 and 92 of his book.

Eisenkolb[693] made 3-year industrial atmospheric-exposure tests of a series of ordinary and copper-bearing steels and commented particularly on the poor showing of the "purer" non-copper samples which, he said, is contrary to general opinion. He called attention to the fact that the oxide impurities actually present in the purer samples are not listed in the ordinary analysis and implied that the poor behavior is due to the presence of oxide. Actually the poor performance of Eisenkolb's "purest" sample can as properly be ascribed to its slightly lower "adventitious" copper content (0.07 per cent) as to its oxygen content. Of three steels of exactly the same copper content (0.11 per cent) the "purer" (0.06 per cent carbon, 0.13 per cent manganese, 0.003 per cent sulphur) gave losses in weight not appreciably different from those of the other two steels with slightly lower carbon, about 0.30 per cent manganese, and 0.02 to 0.03 per cent sulphur.

One is forced to the conclusion that the usual variation in elements (except copper) present in commercial iron and carbon steel will have little effect and that these ferrous materials will all corrode in about the same way, being far more affected by the corrosive environment than by their own composition or structure. While there is doubtless a difference in the way different ordinary unalloyed ferrous materials have acted in actual service, the differences ascribable to the materials themselves fall within such narrow limits that unavoidable differences in service conditions, especially in the initial stage of service, may well overbalance the slight differences ascribable to the materials themselves and alter their order of excellence. Engineering judgment, based on a grasp of the relative magnitude of the different factors in the corrosive environment, and statistical information regarding past performance of various grades of iron and steel under analogous service conditions, may lead to a correct choice among materials so confusingly similar in properties. But accurate prediction of the expected useful life of the materials is still a long way from the exactness of the actuary dealing with the life expectancy of human beings.

Slag inclusions are not ordinarily classed as accelerators of corrosion, though a shattered inclusion which produces a crevice will induce crevice corrosion,*[176] and on a polished surface

* Page 84 of Evans' book.

attack may start about inclusions. However, the inclusions in "dirty steel," while not advantageous for corrosion resistance, cannot be given all the discredit for the corrodibility of steel. Indeed, large amounts of silicate inclusions in wrought iron are given credit by makers of that material as a major cause for the corrosion resistance they ascribe to the iron, on the ground that innumerable slag fibers, once bared, act as a barrier to further corrosion even though the barrier is not a continuous one. Wrought iron differs from low-carbon steels in that its ferrite contains less dissolved impurities other than phosphorus, and because more cementite is present in the low-carbon steels; it differs from ingot iron also by the absence of particles of iron oxide not combined as silicates. It is also lower in manganese, though, according to Hadfield and Friend,[38] approvingly quoted by Evans,[176] manganese is not objectionable from the point of view of corrosion.

200. Corrosion Probability and the Effect of the Common Elements.—In the last few years some interesting work has been done on corrosion probability and conditional velocity of attack. When a piece of iron having a large surface area is exposed to ordinary water, corrosion is practically certain to occur somewhere on this surface. If, however, the exposed area is small—as, for example, when there are small discontinuities or pores in a film of paint—the probability of development of attack is considerably reduced. As has been shown in a number of papers by Evans and Mears and as summarized in a recent paper by Mears,[809] both the probability and the conditional velocity of attack depend upon: (1) variables *external* to the metal, such as the corroding liquid and the surrounding atmosphere, and (2) variables *internal* to the metal, such as the concentration and distribution of minor constituents (carbon, sulphur, and other elements), as well as on the physical condition of the surface and the character of the oxide scale.

External variables (the environment) have, of course, the greatest effect. When, however, the external variables, such as the size of the drops of water used and the composition and gas content of the water, were controlled and when the surface condition of the specimen was standardized, Mears found the following correlation between the composition and the corrosion probability:

Element	Coefficient: Polished with number 00 French emery	Coefficient: Lathe-turned
Carbon	+0.45	+0.26
Sulphur	+0.52	+0.54
Manganese	+0.32	+0.47
Silicon	+0.19	−0.13
Phosphorus	−0.02	−0.20
Copper	−0.12	−0.38

In commenting on these values Mears stated:

> These results indicate that there is a considerable degree of direct or positive correlation between the carbon, sulphur, and manganese content of an iron and its corrosion probability. This means that a material high in any of these constituents, regardless of its composition otherwise, within the limits of the materials tested here, can be expected to have a higher corrosion probability than an iron having a lower carbon, sulphur, or manganese content. Silicon and phosphorus seem to have little or no effect on the probability, while copper has, if anything, a slightly beneficial effect.

According to Mears, additional tests are necessary for a more certain indication concerning the effect of silicon, phosphorus, and copper.

In other words, increasing the amount of carbon, sulphur, or manganese in a steel increases the probability that corrosion will take place if small discontinuities in a paint or other similar surface layer are present. Mears also found that the corrosion probability of scale-covered materials seemed to depend largely or entirely on the physical characteristics of the scale.

201. Influence of Alloying Elements.*—To insure service under corrosive conditions which is of a distinctly higher order than that which can be expected from iron or carbon steel, it is necessary to use alloying elements. The effect of these is brought out in the individual monographs of this series; thus it is only necessary here to sketch in briefest fashion the general effect of some of them.

* The effect of alloying elements has been discussed in detail by Cleaves and Thompson(794) in "The Metal—Iron," pp. 325–328.

Small amounts of copper improve the corrosion resistance of steel in atmospheric service to a marked degree. (See "The Alloys of Iron and Copper,"[705] Chapter IX.)

Large amounts of nickel, 25 per cent or more, make steel resistant to atmospheric corrosion, though invar types, for example, are not always wholly immune. The structural nickel steels with up to 3 per cent nickel are usually considered to be scarcely distinguishable in corrosion resistance from carbon steel. According to Wickenden,* however, 0.25 per cent nickel confers on steel definitely improved resistance to atmospheric corrosion. Low-nickel steels are claimed to resist caustic embrittlement in boilers.

Large amounts of nickel and copper together in cast iron (at least 12.5 per cent nickel and 5 per cent copper) produce a decidedly corrosion-resistant cast iron.[418]

Silicon, at least up to 3 per cent, has relatively little effect on the resistance of steel against the ordinary atmospheric or aqueous corrosion, but the presence of large amounts of silicon in cast iron, about 14 per cent, produces a material which, while very brittle and unmachinable by usual tools, is extremely resistant both to acids and to atmospheric attack.

Chromium is the outstanding alloying element for production of corrosion-resisting ferrous alloys. Rather large amounts, about 11 to 13 per cent, are required to produce truly corrosion-resistant materials—the well-known "stainless steel"—although an improvement in corrosion resistance at high temperature against crude oil is evident with only 4 per cent chromium. The ternary iron-chromium-nickel alloys such as "18-8" and many of that type with higher nickel or chromium or both, used both for corrosion resistance and heat resistance, are now familiar to everyone. These alloys will be dealt with in other monographs of this series.

Vanadium and molybdenum in amounts used in commercial steel are practically without effect, as is tungsten in small quantities. Some slight decrease in corrodibility is thought to result when up to 25 per cent tungsten is added.

Small amounts of aluminum in steel, up to around 0.1 per cent, are without effect. The addition of much larger amounts, say 5 to 10 per cent, has been advocated for heat-resisting pur-

* Private communication.

poses but, while the behavior of such alloys under strongly oxidizing conditions indicates that they might have some value for some corrosion-resistant services, their mechanical properties are as yet so poor that corrosion data have not been obtained. The nitriding steels with around 1 per cent aluminum, 1 per cent chromium, and small amounts of molybdenum are not in themselves corrosion resistant, though the nitrided case is quite resistant to the atmosphere and to fresh and salt water.

Unless we are willing to pay the price for large amounts of chromium or nickel or both, or tolerate the characteristic mechanical properties of the high-silicon alloys, we cannot greatly influence the corrosion resistance of ferrous products. The only alloying element which is both cheap and moderately effective is copper, and, while the copper steels are enough better than plain carbon steels for a limited range of application to be economically desirable, copper does not increase the corrosion resistance sufficiently to produce a material suitable for use in permanent construction without protection. Since the effect of copper appears to consist in the formation of a more impervious and adherent coating of rust rather than in much actual difference in the rate of corrosion of the clean metal surface, the advantages resulting from copper are not realized under conditions where the rust coating is removed by wear from time to time or under those where the conditions for formation of a relatively impervious coating are not met.

The possibilities of control of inherent corrosion resistance of ferrous products by alloying are very limited, and the control is very expensive when alloying is effective. Effective control over the corrosive environment is seldom possible, although in specific cases limited control, such as de-aeration of water, use of inhibitors, or application of external current to make the steel cathodic may be used.

The only general answer to the problem of corrosion of common irons and steels is not to allow the corrosive mediums to touch them. Unless a protective coating is and remains impervious and really prevents contact, it may do more harm than good through localization and intensification of corrosion. In the particular case of soil corrosion of buried pipe, for example, bare pipe may well be more economical for use, save in actually corrosive soils—the so-called "hot spots"—than pipe coated by

methods that do not really delay corrosion enough to be worth the cost.

The great need is for cheap, effective, and permanent protective coatings.

D. OXIDATION AT ELEVATED TEMPERATURES

As a piece of bright iron or steel is heated, it assumes various "temper" colors beginning with a pale-straw color and finally becoming almost black before it becomes "red hot." This phenomenon is attributable to the formation of an oxide coating, and the apparent colors result from interference rather than from the characteristic color of the film. By controlled oxidation at intermediate temperatures dense oxide coatings can be formed which afford considerable protection against certain corrosive environments. At temperatures above a red heat ordinary iron and steel oxidize comparatively rapidly, but the coating is so fragile and non-adherent that it is properly spoken of as scale.

202. Temper Colors.—As pointed out by Gale,[118] the observation of the phenomenon of temper colors is a matter of some antiquity. Three explanations of the cause of temper colors have been advanced: (1) Inherent color of the oxide; (2) interference in the oxide film, as in the case of Newton's rings and soap films; (3) diffraction of incident light by the granular or laminated structure of the film. The second explanation is the most probable.

Gale determined the increase in weight of steel foil resulting from the formation of colored films and calculated the thickness of the films, assuming that their density was 5.25 g. per cu. cm. The thicknesses derived by this method were:

Color	Thickness of Film, cm.
Brownish white	4.2×10^{-7}
Bright brown	6.3×10^{-7}
Dark brown	7.9×10^{-7}
Red brown	12.3×10^{-7}
Purple (?)	17.0×10^{-7}
Dark violet	21.3×10^{-7}

Temper colors depend on both the temperature of the sample and the time that it has been held at temperature but, according to

Tammann,[573] the influence of time is small compared with that of temperature.

203. Protective Coatings.—Under certain controlled conditions of oxidation, adherent oxide coatings may be formed on iron or steel. A good discussion of these coatings will be found in the books by Rawdon[281] and Hedges.[529]

In the Bower-Barff process, articles are heated to about 540°C. (1000°F.), superheated steam is admitted into the furnace for about 30 min. and then replaced by a hot reducing gas, and the cycle repeated until a coating of the desired thickness has been produced.

Tenacious oxide coatings can be formed by simply heating bright iron in air or in a bath of fused nitrates, but, according to Rawdon, these coatings are ornamental rather than protective.

Oxide coatings resulting during casting, hot working, or annealing afford some protection to the underlying metal, and the character of the coating produced can be controlled to a limited extent.

204. Rate of Scaling.—At temperatures above 500 or 600°C. (930 or 1110°F.) ordinary iron or steel oxidizes comparatively rapidly, and the oxide coating formed does not protect the underlying metal from further oxidation. When ordinary iron or steel is held for a long period of time at high temperature, it is gradually converted to oxide and may entirely waste away much as it wastes away when exposed to particularly corrosive conditions at ordinary temperature. For effective resistance to scaling at high temperatures it is necessary to use an alloy steel or iron just as it is necessary to use alloyed materials for effective resistance to ordinary-temperature corrosion.

Data given recently by MacQuigg[471] illustrate the rapid increase in rate of oxidation of plain carbon steels, containing 0.15 to 0.20 per cent carbon, when exposed for one week to an oxidizing atmosphere at increasingly higher temperatures. The specimens used were cubes approximately 0.5 in. on a side. After the given exposure, the adherent scale was flaked off as completely as possible without removing appreciable metal. The cubes were weighed after each period of heating, and the loss was calculated and expressed as a percentage of the original weight. Results were as follows:

Temperature		Loss in weight in 1 week, per cent
°C.	°F.	
300	570	0.09
400	750	0.22
475	890	0.52
550	1020	0.43
650	1200	3.33
750	1380	15.10
850	1560	36.80
1000	1830	100.00

From considerations of the mechanism of scale formation and from experimental data, Pilling and Bedworth(105) concluded that the rate of oxidation of iron at a constant (high) temperature can be expressed by the following equation:

$$\frac{dW}{dt} = k'\left(\frac{1}{W}\right)$$

where W is quantity of oxygen, t time, and k' a constant. By integrating they obtained the equation:

$$W^2 = kt$$

where k is a new constant. According to this equation, the amount of oxide formed at a constant temperature varies directly as the square root of the elapsed time.

Heindlhofer and Larsen(714) presented experimental data on the rate of oxidation of commercially pure iron at various constant temperatures (Fig. 194) which show that the above relationship is fundamentally correct, though it may be modified by conditions such as blistering of the scale which disturbs the growth of a continuous sheet of oxide. These authors discussed the chemical and physical properties of the oxides of alloying elements for steel in respect to the modification in scaling properties which they produce and explained why rather large amounts of readily oxidizable elements are required to form a protective scale and thus produce what is ordinarily called a non-oxidizable alloy, but which is actually corrosion resistant because it is oxidizable.

Since blistering of the scale is one of the disturbing factors in the rate of oxidation, and since blistering is a nuisance in

producing steel of smooth finish, Griffiths[706] studied the factors involved in making steel with a uniform layer of scale. He concluded that an inert gas, such as the nitrogen of the air, may diffuse through the scale and, where local impurities or inclusions keep the scale from adhering tightly, this accumulation tends to lift the scale into a blister. He stated that blisters do not form readily in an atmosphere of pure oxygen. In experi-

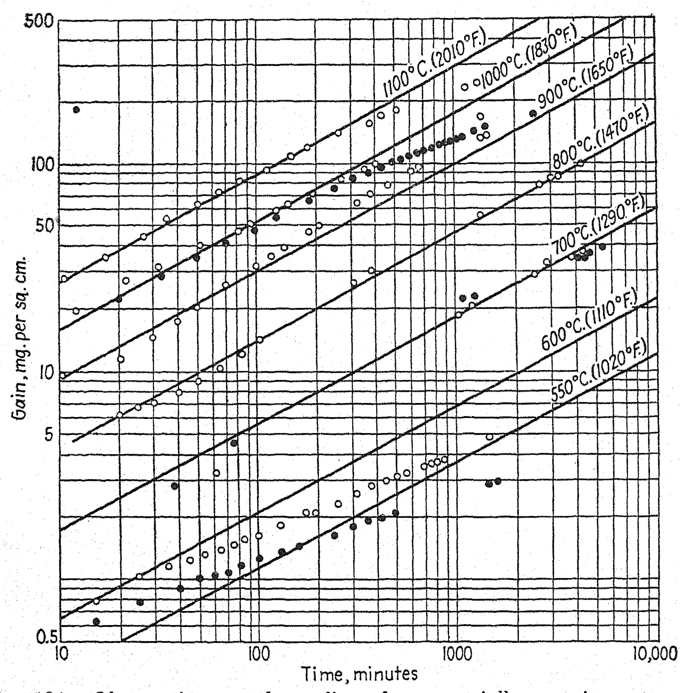

FIG. 194.—Observations on the scaling of commercially pure iron at various constant temperatures between 500 and 1100°C. (1020 and 2010°F.). The lines are drawn with the slope required by the theory that the rate of scaling is proportional to the square root of the elapsed time. (*Heindlhofer and Larsen.*[714])

menting, he found that steam added to the furnace atmosphere in the range of 850 to 1000°C. (1560 to 1830°F.), within which blistering in air is most serious, remedied blistering. His explanation is that the surface is more slowly heated so that the initial layer of scale is not disrupted at the start, since very slow heating in air also produces less blistering. This explanation does not seem very satisfying. He claimed that in the

grain boundaries and gives the scale a tenacious hold, while changes in carbon content do not affect the adherence.

The scaling of steel has been studied also at the University of Michigan[477,643,668,669] where various atmospheres, steels, temperatures, and times of exposure were used, and the amount of scale formed was determined by weighing the descaled sample or by actually weighing the scale. Below 705°C. (1300°F.) there was little difference between the scaling action of steam, carbon dioxide, oxygen, and air, but between 705 and 915°C. (1300 and 1680°F.) steam was the most active oxidizing agent. From 915 to 1080°C. (1680 to 1980°F.) oxygen was the most active, and above 1080°C. (1980°F.) steam was again the most active. Air formed more scale than steam in the region 940 to 1010°C. (1720 to 1850°F.). Carbon dioxide was for the most part less active than the other oxidizing gases. For gas-air mixtures at temperatures up to 925°C. (1700°F.) an increase in carbon monoxide from nil to 4 per cent decreased the amount of scale formed by at least 50 per cent. Increasing the amount of carbon monoxide from 4 to 17 per cent did not greatly decrease the amount of scale formed. At higher temperatures, increasing the carbon monoxide content from nil to 12 per cent gradually decreased the amount of scale formed. The amount of scale formed did not have a simple relation to the carbon content of the steel. Scaling losses for several commercial steels at temperatures of 925 and 1260°C. (1700 and 2300°F.) are shown in Fig. 195. These are for relatively short times and represent the initial behavior rather than that for long periods of exposure. The data from which the curves were plotted were obtained from samples heated in mixtures of air and commercial gas, the ratio of gas to air being varied to produce different atmospheres. These curves show that as the temperature is increased the carbon monoxide concentration must be greatly increased if severe scaling is to be prevented.

Another investigation was reported by Schroeder,[569] who determined the scaling losses of a 0.34 per cent carbon steel heated to various temperatures in atmospheres encountered in industrial practice. He found that for heating in an oxidizing atmosphere the scaling increased by 50 per cent as the temperature was increased from 900 to 950°C. (1650 to 1740°F.) and by 100 per cent as the temperature was increased from 900 to 1000°C. (1650 to 1830°F.).

205. Composition of Scale.—Careful studies of the structure and composition of scale were reported by Pfeil,[329,479] who found that it normally consists of three distinct layers differing in structure and composition. The layers consist of ferric oxide, magnetic oxide, and a decomposed ferrous phase. The two outer layers, of ferric oxide and magnetic oxide, adhere firmly to one another, but the magnetic oxide tends to break away from the decomposed ferrous phase, which is, of course, the one in contact with the metal.

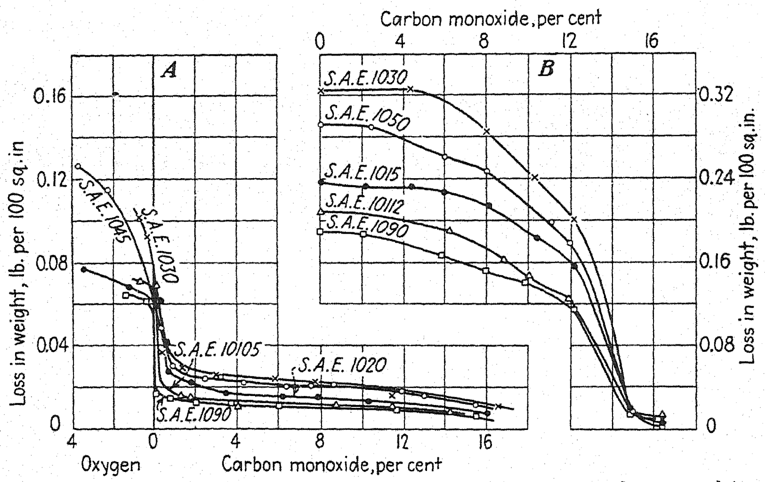

FIG. 195.—Scaling losses of some commercial steels in atmospheres resulting from the combustion of air and commercial gas. The loss in weight was determined after (*A*) 60 min. at 925°C. (1700°F.) and (*B*) 40 min. at 1260°C. (2300°F.). (*Upthegrove.*[668])

In a recent paper Endo[796] reported the results of an investigation on amount of scaling and on the composition of the scale formed. By means of a thermobalance he determined the increased weight by oxidation for 20 to 60 min. at 800 to 1200°C. (1470 to 2190°F.) of six steels containing 0.11 to 0.62 per cent carbon. Two formulas for scaling were derived, one for "mild" steel and one for higher carbon materials. From X-ray and microscopic examinations the scale was found to consist of an outer layer of Fe_2O_3, a central layer of Fe_3O_4, and an inner portion, adjacent to the unoxidized steel, which consisted of a mixture of Fe_3O_4 and FeO.

206. Decarburization.—In many heat-treating operations it is necessary to take special precautions to prevent scaling or

decarburization, often at a considerable expense. Sheets, and other shapes, are sometimes annealed in a reducing atmosphere in order to prevent discoloration or scaling. Tools and other small articles may be prevented from scaling during heat treatment by heating in contact with a carbonaceous material, by heating in a salt bath, or by heating in a controlled atmosphere.

Heat-treating atmospheres which prevent scaling do not necessarily prevent decarburization; in fact, atmospheres that cause some scaling may produce less decarburization than others that completely prevent scaling. An article by Curran and Williams,[252] which contains a good bibliography on decarburization, cites results which indicate that the surface of high-carbon steels is decarburized by heating in an atmosphere that is apparently strongly reducing. Decarburized surfaces were found on samples packed in charcoal and heated, but addition of sodium carbonate to the charcoal tended to prevent decarburization. Jominy[461] studied the decarburizing action of various gas-air atmospheres on high-carbon steels and found that, given sufficient time, even atmospheres containing an appreciable amount of carbon monoxide produce surface decarburization. Tour[575] published a paper on the use of controlled atmospheres in heat treating and gave the composition of so-called neutral atmospheres. According to him, the formation of a soft skin on carbon tool steels during heat treatment can be prevented by the use of a "combusted" fuel-gas atmosphere containing a small amount of free oxygen.

According to the information available, a given atmosphere may cause surface decarburization of certain steels but not of others. It is, therefore, not possible to set up rigid specifications for a neutral atmosphere, and a really satisfactory knowledge of the reactions between atmospheres and high-carbon steels can be gained only by further investigation.

E. AUTHOR'S SUMMARY

1. Corrosive attack on iron or steel objects usually results in a gradual and uniform wasting away of the object unless it is exposed to an environment in which the attack is localized. Except in rare instances, intergranular penetration and brittleness do not result during or from the rusting of unalloyed steel or iron.

2. Corrosion resistance cannot be expressed in absolute units.

3. Accelerated tests using an acid or other rapidly acting reagent as the corroding medium are of little use in evaluating the corrosion resistance of iron and steel. Despite the claims made in former years, acid-solution tests do not indicate the relative resistance of irons and steels to atmospheric corrosion.

4. The probable rate of corrosion or rusting of a material in any environment can generally be estimated only from long-time tests made in substantially the same environment.

5. It cannot be stated with certainty that wrought iron is more resistant to corrosive attack than ingot iron, or *vice versa*. Neither is it certain that increased carbon content of a steel decreases resistance to corrosion. The corrosion resistance of cast iron compares favorably with that of ingot iron and wrought iron.

6. Variations in the content of ordinary impurities (save copper) of iron or steel have little influence on corrosion resistance. Opinion is divided on the effect of oxygen.

7. To increase the corrosion resistance of iron or steel greatly, it is necessary to resort to alloying. Small amounts of copper markedly improve resistance to atmospheric attack. Larger amounts of other elements such as chromium, silicon, or nickel greatly improve corrosion resistance; the effects of these, and other elements, are more properly discussed in the other monographs of this series.

8. Temper colors result from interference caused by thin oxide layers; the apparent color is dependent on the thickness of the layer and is probably not a function of the composition of the layer.

9. Some oxide coatings increase appreciably the resistance to corrosion.

10. The rate of scale formation at elevated temperatures is dependent on the composition of the surrounding atmosphere and on the temperature. In an oxidizing atmosphere the rate of scale formation increases rapidly as the temperature increases above 900°C. (1650°F.).

11. A reducing atmosphere can produce a decarburized surface on high-carbon steel. Heat-treating atmospheres must be carefully controlled if scaling and decarburization are to be prevented.

CHAPTER XV

MISCELLANEOUS PHYSICAL PROPERTIES OF IRON-CARBON ALLOYS

Elastic Constants—Density and Related Properties—Thermal Expansion—Other Properties—Author's Summary

This chapter deals with certain physical properties which have been more or less carefully investigated, often as a step in the study of theories of matter. An exact knowledge of these properties may or may not be required in the task of choosing a material for a particular use in engineering. When exact data are required, it is normally necessary to determine them upon the material to be used, since calculation and extrapolation from available data are seldom entirely safe in the case of an industrial ferrous alloy whose composition, when all impurities are considered, may be complex and whose structure depends on the working and heat treatment.

When a rough idea of the properties is all that is required, a knowledge of general trends in the case of high-purity iron, on which many of the determinations have been made, and a qualitative idea of the effect of carbon, may suffice.

The agreement among the physical constants determined by various observers is on the whole not so good in the case of iron and iron-carbon alloys as it is for many other metals. It must be concluded that the physical properties of ferrous materials are especially sensitive to variation in composition and structure.

Considering commercial iron-carbon alloys as pure iron to which additions of carbon and other elements have been made, the properties of these alloys can be, in part, predicted and compared in terms of the effects which these additions have on the properties of pure iron. Pure iron is hence a useful base-line, particularly as its properties may be presumed to be better known than those of any alloy. Thus, in the following discussion, it was often necessary to refer briefly to the properties of iron itself, which are set forth in detail in "The Metal—Iron."[794]

A. ELASTIC CONSTANTS

The elastic constants of steel are fundamental properties and, as such, are useful to the designing engineer. Inspection of the data available on these properties leads to the conclusion that variations of magnitude exist for which there is no explanation. For example, taking into consideration composition and structure leads to little useful information. Approximate or average values seem to be adequate for most engineering applications, consequently there has been small incentive to undertake precise measurements. It seems that, for the present at least, the metallurgist can do almost nothing to influence the numerical value of these properties.

207. Modulus of Elasticity.—The (Young's) modulus of elasticity, which numerically is tension divided by elongation per unit length, and which may be considered a measure of the stiffness of a material, is essentially constant for different steels. Reported values range from 27 to 31 million lb. per sq. in. The figure is often rounded to 30 million, but, according to Laurson and Cox,[467] 29 million is a more exact value.

It is often alleged that some alloy steel or other in a particular condition of heat treatment has a greater modulus of elasticity than usual; therefore, that it is a better spring material. Determination of the modulus of elasticity demands careful manipulation and an accurate extensometer; difference of values obtained by different observers is ascribable more often to difference of precision or of testing procedure than to difference of the steels. Furthermore, the stored energy—resilience—is of more interest to the spring maker than is modulus of elasticity.

For iron-carbon alloys, Honda and Tanaka[180] found the modulus of elasticity to decline from a value of 29.7 million for annealed low-carbon iron to 27.5 million lb. per sq. in. for an annealed 1.48 per cent carbon alloy. For quenched alloys, smaller values were found: *e.g.*, 26.75 to 28.5 million for 0.10 to 1.19 per cent carbon alloys and 25.9 million for the 1.48 per cent carbon alloy. Moore,[327] however, stated that the modulus of elasticity is not changed by heat treatment. This statement may be supported by the inconclusive results of Lyse and Godfrey,[634] who found values ranging from 27.5 to 30.5 million lb. per sq. in. for carbon and alloy steels in either the annealed or the quenched

and tempered condition. No consistent difference ascribable to heat treatment was found.

Probably the most reliable data on the modulus of elasticity of carbon steels were reported by Abram.[676] Unusual care was exercised in technique and in preparation of the specimens. Results are given in Table 94. It is to be seen that the modulus of elasticity of Armco iron and mild steel is definitely greater than that of medium- and high-carbon steels, although the exact effect of carbon cannot be determined because of variation of manganese content. Abram concluded provisionally that the modulus of iron in the annealed condition is about 30.2 million lb. per sq.

TABLE 94.—MODULUS OF ELASTICITY OF CARBON STEELS*

Material	Approximate composition, per cent		Modulus of elasticity, million lb. per sq. in.
	C	Mn	
Armco iron	0.03	0.03	30.1
Mild steel	0.16	0.58	30.11
High-tensile steel	0.31	1.34	30.03
Carbon steel	0.33	0.72	29.92
Carbon steel	0.66	0.56	29.67
Carbon steel	0.88	0.88	29.79
Carbon steel	1.04	0.38	29.73

* Abram.[676]

in.; that, with increasing carbon content, the value declines to about 29.7 million; and that the modulus of medium- and high-carbon steels is approximately constant, although increased manganese content tends to increase its value.*

Higher values of modulus of elasticity for cold-worked than for annealed steel have been reported, but, according to Jeffries and Archer,[125] this is an error arising out of the fact that moduli for annealed metal were computed for stresses exceeding the elastic limit.

Although austenite and ferrite differ in crystal structure, the moduli of elasticity of austenitic and ferritic steels have been reported to lie within the same range. Thum[665] quoted 29 mil-

* The modulus of elasticity is affected by temperature. This is discussed in section 161, p. 438.

lion lb. per sq. in. for 18 per cent chromium, 8 per cent nickel steel, and Monypenny[475] gave the range 26.5 to 30 million for the same material; these values will be seen to be indistinguishable from those given for carbon steels.

The definition of modulus of elasticity demands that strain is proportional to stress. For such materials as cast iron (and possibly 18 per cent chromium, 8 per cent nickel steel) this is not true; for them an approximation called the "secant modulus" is made. The secant modulus is defined as the slope of the straight line connecting a given point on the stress-strain curve with the origin. According to Salmon,[487] the secant modulus for gray cast iron ranges from 12 to 20 million lb. per sq. in., and the values for white cast iron range from 20 to 25 million. The value for malleable iron is about 25 million. As may be expected, the value of the modulus of elasticity increases as the amount of free graphite decreases.*

208. Poisson's Ratio.—When a specimen is elastically stressed in tension it becomes smaller in cross-section. If the volume would not change, the ratio (called "Poisson's ratio") of lateral contraction per unit width to increase in length per unit length would be 0.5. Actually, the volume increases, so that this ratio is not 0.5 but, for steel, is more nearly 0.3.

The precise determination of the small change in diameter of a tensile bar is even more difficult than that of change of length, consequently special apparatus and great care are required to secure reliable data.

Jasper[124] calculated Poisson's ratio from elastic constants, using the modulus of elasticity and the modulus of rigidity as determined by torsion tests (see page 567), and made direct determinations of the ratio in the tensile test as well. His results at 16°C. (60°F.) follow:

Carbon, per cent	Condition	Poisson's ratio	
		Observed	Calculated
0.02	Rolled	0.197	0.188
0.41	Normalized	0.241	0.245
0.90	Normalized	0.241	0.243
1.20	Normalized	0.260	0.241

* The modulus of elasticity of cast iron is discussed in more detail on p. 291. Stress-strain curves for gray iron are illustrated on pp. 292 to 298.

Several alloy steels yielded values from 0.26 to 0.27. The moduli of elasticity of all the steels at 16°C. (60°F.) ranged from 29.9 to 30.6 million lb. per sq. in., while the torsional moduli ranged from 11.8 to 12.6 million.

Jasper determined also the variation with temperature of Poisson's ratio for 0.49 per cent carbon steel in the normalized condition. Results were: 0.267 at −19°C. (0°F.), 0.245 at 16°C. (60°F.), and 0.231 at 38°C. (100°F.). Keulegan and Houseman[(626)] found Poisson's ratio to remain almost constant with change of temperature.

Jasper's values are smaller than those usually given for steel. Lyse and Godfrey[(634)] found the range 0.272 to 0.320 for annealed and for quenched and tempered carbon and alloy steels, and 0.272 to 0.302 for rolled carbon structural steels. They cited several references to the literature showing that the value for structural steel ranges from 0.27 to 0.30. In discussion, H. F. Moore reported that he found Poisson's ratio on a 10-in.-diameter steel shaft, in compression, applied by a 3 million-lb. testing machine, to be 0.30 to 0.32. According to Jasper, cold-rolled steel in tension has a relatively small ratio.

The variation of Poisson's ratio is of the same order as that of the elastic modulus, and likewise cannot be certainly associated with variations in composition and heat treatment. Nor is there definite indication of what, if any, correlation there may be between these values and those of other properties of the steels. It has been suggested,[(124)] on the basis of the determination of Poisson's ratio of two rails of which one had transverse fissures, that such determinations might throw light on the fissure problem, but this has not been substantiated.

209. Modulus of Rigidity.—The modulus of rigidity, sometimes called the torsional modulus, is determined from values of shear stress and strain. It is related to the modulus of elasticity by:

$$G = \frac{E}{2(1 + \sigma)}$$

where G is the modulus of rigidity, E is the modulus of elasticity, and σ is Poisson's ratio.

Zimmerli, Wood, and Wilson[(426)] found a range of values at room temperature of 10.7 to 11.2 million lb. per sq. in. for the

modulus of rigidity of hard-drawn 0.65 per cent carbon steel, for patented wire containing 0.71 to 0.90 per cent carbon, and for severely cold-worked music wire. Duplicate tests of annealed music wire yielded values of 10.3 and 11.7 million, as compared with 11.2 million lb. per sq. in. for the cold-worked wire. These workers concluded—for a reason which is not evident—that modulus of rigidity is increased in value by cold work; Jasper, in discussion, disagreed with this conclusion. Results of Zimmerli, Wood, and Wilson on the effect of temperature on the modulus of rigidity were inconclusive, but Jasper (again in discussion) quoted eclectic values of 12+ million at −100°C. (−148°F.), 11.7 million at room temperature, and about 9.5 million at 500°C. (930°F.).

210. Bulk Modulus—Compressibility.—The bulk modulus is the ratio of pressure (applied equally in all directions) to the resulting fractional decrease of volume of a body. It is related to modulus of elasticity by

$$K = \frac{E}{3(1 - 2\sigma)}$$

where K is the bulk modulus, E is the modulus of elasticity, and σ is Poisson's ratio.

According to Bridgman,[81] the variation of volume of iron with pressure (0 to 12,000 kg. per sq. cm.) can be expressed by:

$$\frac{-\Delta V}{V_0} = 5.87 \times 10^{-6}p - 2.1 \times 10^{-12}p^2 \qquad \text{at } 30°\text{C. } (85°\text{F.})$$

and

$$\frac{-\Delta V}{V_0} = 5.93 \times 10^{-6}p - 2.1 \times 10^{-12}p^2 \qquad \text{at } 75°\text{C. } (165°\text{F.})$$

where p is in kg. per sq. cm.

There are no data on steels; however, no appreciable change in the value of bulk modulus with carbon content is to be expected, because, even though there is some variation of the modulus of elasticity, Poisson's ratio also varies.

B. DENSITY AND RELATED PROPERTIES

Density and its reciprocal, specific volume, are closely related to thermal expansion and compressibility. Thermal expansion

may be considered as the change in volume due to change in temperature, and compressibility the change in volume due to change in pressure.

As a practical datum the density of pure iron calculated from direct determination, extrapolation, and lattice constant may be taken as 7.87 g. per cu. cm. at 20°C. (70°F.), which corresponds to a specific volume of 0.1271 cu. cm. per g.

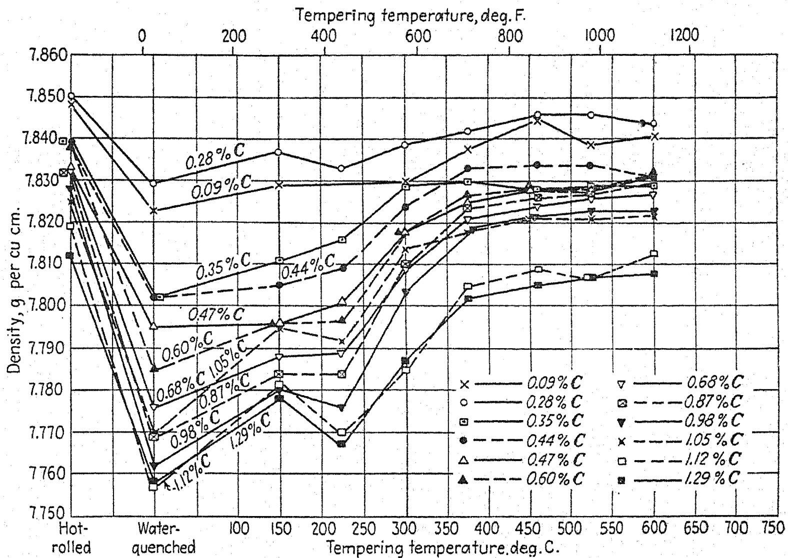

FIG. 196.—Density of carbon steels in the hot-rolled, quenched, and quenched and tempered conditions. (*Cross and Hill.*[208])

211. Effect of Impurity on the Density of Iron.—For small additions to iron the change in density may be considered proportional to the amount of added element. The results of Benedicks[7] are frequently cited. Gumlich[47] and others have also investigated the effect of various additions. The agreement among various investigators is poor, and the density assumed for zero impurities differs so much that the results are of limited usefulness.

212. Density of Carbon Steels.—A careful study of the density of plain carbon steels was made at the National Bureau of Standards by Cross and Hill.[208] The three purest samples of iron examined by them gave an average of 7.867 g. per cu. cm.

These results are shown in Fig. 196. The density in the hot-rolled condition could be represented by:

$$\text{Density} = 7.855 - 0.032C$$

where C is the percentage of carbon. On annealing, they found:

$$\text{Density} = 7.860 - 0.04C$$

Quenching decreased the density, the magnitude of this decrease rising with increased carbon content up to 1 per cent carbon.

Ralston[333] averaged the available data on the density of the metallographic constituents as follows:

Constituent	Specific volume, cu. cm. per g.	Density, g. per cu. cm.
Iron	0.1271	7.864
Cementite	0.1304	7.67
Pearlite	0.1286	7.778
Austenite with 0.9 % C	0.1275	7.843
Martensite with 0.9 % C	0.1310	7.633
Alpha martensite with 0.9 % C	0.1319	7.581
Beta martensite with 0.9 % C	0.1282	7.800

It is well known that cold work decreases the density of non-porous iron or steel. O'Neil[132] and Ishigaki[183] both found that Armco-iron tensile samples decreased in density in the necked-down portion, the decrease being from a density of about 7.87 to 7.84 where the reduction of area was about 50 per cent. Annealing restored the density to the original value.

The density of iron-carbon alloys, especially at elevated temperatures, has been studied by Tammann and Bandel,[778] who gave the data reported as specific volume, shown in Fig. 197.

213. Density of Molten Iron Alloys.—Benedicks, Ericsson, and Ericson[287] investigated the specific volume of a series of alloys in the molten condition, ranging in carbon content from 0.03 to 4.4 per cent. Table 95 shows their results. Figure 198 shows the effect on specific volume at 1600°C. (2910°F.) of additions of carbon, aluminum, silicon, phosphorus, chromium, manganese, nickel, and tungsten.

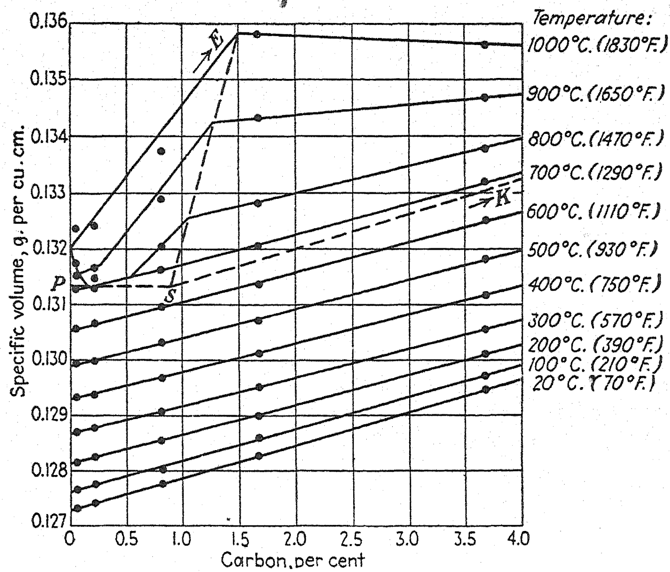

FIG. 197.—Specific volume of iron-carbon alloys at various temperatures. (*Tammann and Bandel.*[778])

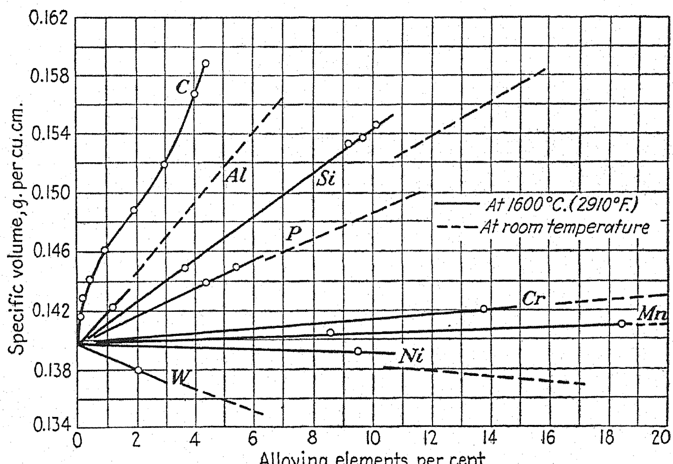

FIG. 198.—Effect of various alloying elements on specific volume of iron-carbon alloys at 1600°C. (2910°F.) and at room temperature. (*Benedicks, Ericsson, and Ericson.*[287])

214. Volume Change on Melting.—High-purity iron shows the normal increase in volume on melting. The specific volume of solid iron at 1530°C. (2785°F.), according to the X-ray measurements of Schmidt,(654) is 0.137 cu. cm. per g., and Benedicks and associates'(287) value for a molten iron at 1530°C. (2785°F.) is 0.138 cu. cm. per g., indicating an increase in volume of about 1 per cent.

Sauerwald(235) found that white cast iron (4.15 per cent carbon, 2.2 per cent manganese) increased in volume 1.4 per cent on melting, whereas gray iron (3.32 per cent carbon, 2.76 per cent

TABLE 95.—AVERAGED VALUES FOR THE SPECIFIC VOLUME AND DENSITY OF MOLTEN IRON-CARBON ALLOYS AT THEIR MELTING POINT AND AT 1600°C. (2910°F.); ALSO THE CHANGE IN SPECIFIC VOLUME PER 100°C. (180°F.)*

Carbon content, per cent	Melting point, °C.	Specific volume, cu. cm. per g.		Change in specific volume per 100°C. (180°F.)	Density, g. per cu. cm., 1600°C. (2910°F.)
		At melting point	At 1600°C. (2910°F.)		
0.0	1533	0.1384	0.1397	0.0020	7.158
0.1	1514	0.1399	0.1416	0.0021	7.061
0.2	1503	0.1407	0.1428	0.0021	7.003
0.3	1494	0.1412	0.1436	0.0022	6.963
0.4	1486	0.1416	0.1441	0.0022	6.939
0.5	1480	0.1419	0.1445	0.0023	6.920
0.6	1477	0.1421	0.1448	0.0023	6.905
0.7	1474	0.1423	0.1451	0.0023	6.891
0.8	1469	0.1424	0.1454	0.0024	6.877
0.9	1464	0.1425	0.1457	0.0024	6.863
1.0	1458	0.1425	0.1461	0.0025	6.844
1.5	1422	0.1423	0.1471	0.0028	6.798
2.0	1382	0.1423	0.1487	0.0030	6.725
2.5	1341	0.1417	0.1501	0.0033	6.662
3.0	1290	0.1413	0.1518	0.0035	6.587
3.5	1232	0.1409	0.1539	0.0037	6.499
4.0	1170	0.1403	0.1566	0.0038	6.385
4.4	1190	0.1430	0.1587	0.0038	6.301

* Benedicks, Ericsson, and Ericson.(287)

silicon, 0.49 per cent phosphorus) decreased 0.6 per cent on melting. The abnormal decrease in the gray iron was ascribed to the volume contraction resulting from the formation of cementite. As discussed on pages 264 and 265, Bardenheuer and Ebbefeld(148) and Bardenheuer and Bottenberg(434) concluded that evolution of gas was a factor in the expansion of gray iron during solidification. As a general rule, iron-carbon alloys expand on melting and contract on solidification. During solidification the evolution of gas and the formation of graphite may lead to an apparent increase in volume.

C. THERMAL EXPANSION

Density and thermal expansion are related by the expression:

$$D_t = \frac{D_0}{1 + 3\alpha t}$$

where D_t is the density at temperature t, D_0 is the density at temperature zero, and α is the coefficient of linear thermal expansion. This relation is valid only for isotropic crystalline substances for which the coefficient of cubical expansion may be considered to be equal to 3 times the linear coefficient. The expression is not useful for determination of expansion; it is useful rather for computation of density at elevated temperature, since the temperature range is narrow over which experimental determination of density is feasible. An even better method, perhaps, is computation of (ideal) density from X-ray measurements.

Two different coefficients are reported in the literature: (1) the "true" coefficient, which is given by the slope of the expansion-temperature curve at any point, and (2) the "mean" coefficient, which is the average expansion per degree over a definite temperature range.

215. Thermal Expansion of Steels and Cast Irons.—Most data on thermal expansion have been obtained by direct measurement. As a basis of comparison, mean coefficients of expansion of iron are given in Table 96. (Additional data may be found in "The Metal—Iron."(794))

The influence of addition of other elements to iron on the mean coefficient of expansion between 20 and 450°C. is shown in Fig. 199 as given by Maurer and Schmidt.(75) Mochel(474) summarized the data on coefficients of expansion of iron-carbon

TABLE 96.—MEAN COEFFICIENTS OF EXPANSION OF IRON

Austin and Pierce*(679)		Schmidt†(654)		Esser and Müller‡(608)	
Temperature range, °C.	Coefficient $\alpha_m \times 10^6$	Temperature range, °C.	Coefficient $\alpha_m \times 10^6$	Temperature range, °C.	Coefficient $\alpha_m \times 10^6$
0 to 100	11.9	20 to 100	11.6	20 to 100	13.5
0 to 400	13.7	20 to 400	12.0	20 to 400	14.8
0 to 700	14.9	20 to 740	12.7	20 to 750	16.2
0 to 800	14.9	20 to 900	13.6	20 to 900	15.7
		20 to 1400	16.2		
		910 to 1400	24.1	900 to 1100	24.0

* Selected average.
† Determined by X-ray measurements on electrolytic iron containing 0.01 per cent carbon.
‡ Determined by X-ray measurements on carbonyl iron.

alloys. The expansion of steel is in general somewhat less than that of high-purity iron; that of cast iron may be considerably less,

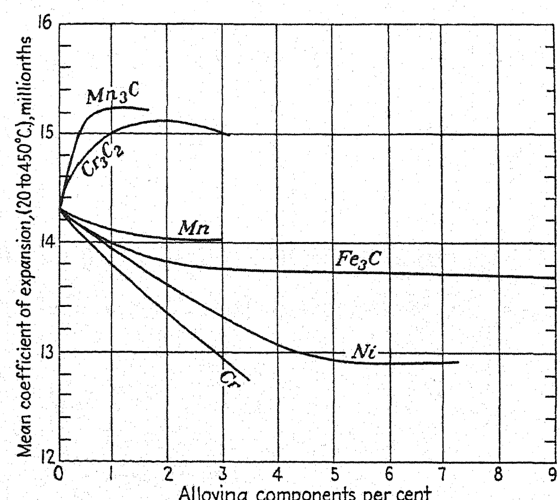

FIG. 199.—Effect of various elements on the mean coefficient of expansion of iron. (*Maurer and Schmidt.*(75))

certain annealed samples giving a coefficient of expansion as low as 8.6×10^{-6} per °C. Mochel's tabulation for annealed, or normalized and tempered, steels shows that the coefficient is not markedly affected by carbon content. The coefficient from 0 to

TABLE 97.—MEAN COEFFICIENT OF EXPANSION OF IRON-CARBON ALLOYS*

Temperature interval, °C.	Carbon, per cent														
	0.06	0.09	0.22	0.33	0.40	0.56	0.65	0.81	1.25	1.45	1.67	1.97	2.24	3.66	3.80
	Mean coefficient of expansion × 10^6														
20 to 50	11.49	11.16	11.22	10.92	10.73	10.17	10.74	10.17	10.58	9.67	10.02	9.72	9.03	8.57	8.43
20 to 100	11.66	11.58	11.66	11.09	11.29	10.98	11.04	11.04	10.87	10.13	10.44	9.94	9.61	8.59	8.71
20 to 150	12.06	12.06	11.96	11.70	11.45	11.44	11.34	11.43	11.14	10.38	10.51	10.02	.....	8.75	8.76
20 to 200	12.32	12.61	12.12	11.89	11.99	11.85	11.57	11.56	11.08	10.58	10.28	9.96	9.64	8.83	8.49
20 to 250	12.56	12.72	12.49	12.38	12.07	12.27	11.88	11.93	11.46	11.16	10.83	10.52	10.25	9.04	9.35
20 to 300	13.02	13.01	12.78	12.72	12.47	12.65	12.31	12.43	11.87	11.67	11.38	11.14	10.96	9.88	10.11
20 to 350	13.34	13.36	13.12	13.09	12.86	13.08	12.74	12.84	12.33	12.18	11.88	11.72	11.59	10.63	10.92
20 to 400	13.65	13.63	13.38	13.42	13.26	13.40	13.16	13.21	12.75	12.68	12.34	12.21	12.15	11.35	11.52
20 to 450	13.98	13.93	13.68	13.88	13.66	13.74	13.42	13.55	13.16	13.13	12.82	12.68	12.72	11.96	12.12
20 to 500	14.22	14.18	13.93	14.02	13.90	14.02	13.84	13.82	13.36	13.48	13.15	13.12	13.16	12.50	12.62
20 to 550	14.33	14.38	14.17	14.20	14.17	14.27	13.93	14.08	13.78	13.78	13.38	.....	.....	12.91	
20 to 600	14.64	14.64	14.38	14.43	14.36	14.50	14.20	14.22	13.99	14.04	13.66	.....	.....	13.23	
20 to 650	14.85	14.86	14.66	14.59	14.61	14.67	14.52	14.50	14.22	14.24	13.93	.....	.....	13.70	
20 to 700	15.01	15.03	14.81	14.76	14.76	14.81	14.65	14.67	14.41	14.42	14.24	.....	.....	13.92	
20 to 750	14.84	14.93													
20 to 800	14.67	14.61	12.93	11.33	11.72	12.46	12.68	14.22	14.78	15.13	15.02	.....	.....	14.13	
20 to 900	13.14	12.34	12.48	11.53	12.35	13.54	13.87	15.05	16.64	17.63	17.57	.....	.....	15.04	
20 to 1000	13.35	13.32	13.16	13.09	13.25	14.38	14.76	15.70	17.43	19.50	19.61	.....	.....	15.86	
700 to 1000	9.50	9.42	9.40	9.29	9.91	13.47	15.00	17.90	24.10	29.74	31.45	.....	.....	20.24	

* Driesen.[28]

TABLE 98.—MEAN COEFFICIENT OF EXPANSION OF IRON-CARBON ALLOYS*

Temperature interval, °C.	Carbon, per cent															
	0.06	0.09	0.22	0.33	0.40	0.56	0.65	0.81	1.08	1.25	1.45	1.67	1.97	2.24	3.66	3.80
	Mean coefficient of expansion × 10^6															
20 to 100	11.66	11.58	11.66	11.08	11.29	10.98	11.04	11.04	10.77	10.87	10.13	10.44	9.94	9.61	8.58	8.71
100 to 200	12.93	12.93	12.52	12.54	12.35	12.38	12.03	11.94	11.56	11.24	11.00	10.12	9.97	9.67	7.94	7.93
200 to 300	14.23	14.23	13.96	14.13	13.74	14.10	13.58	14.00	13.47	13.28	13.66	13.33	13.25	13.36	12.77	12.98
300 to 400	15.34	15.39	15.06	15.34	15.46	15.35	15.16	15.13	15.13	15.10	15.41	15.13	15.19	15.41	15.43	15.48
400 to 500	16.27	16.22	15.92	16.18	16.22	16.26	15.96	16.22	16.48	16.08	16.46	15.55	16.44	16.97	16.79	16.79
500 to 600	16.64	16.28	16.47	16.41	16.47	17.24	16.55	16.73	.....	16.43	16.64	16.43	.....	.....	16.90	
600 to 700	17.03	17.14	17.03	16.47	16.85	16.48	17.13	16.45	.....	16.71	16.55	17.28	.....	.....	17.68	
700 to 800	+12.12	+11.59	+ 0.08	−11.76	− 8.71	− 3.46	− 0.74	+10.97	.....	+17.08	+19.85	+20.14	.....	.....	+15.34	
800 to 900	− 5.89	− 5.30	+ 8.82	+12.87	+17.09	+21.66	+22.95	+21.69	.....	+31.05	+36.74	+37.20	.....	.....	+21.98	
900 to 1000	+22.18	+21.65	+21.00	+21.00	+20.86	+21.79	+21.38	+21.04	.....	+23.90	+31.50	+37.10	.....	.....	+22.72	

* Driesen.[28]

100°C. varies from 11.1 to 12.8 for carbon contents from 0.006 to 0.94 per cent. That for 600 to 700°C. is given as 17.4 for 0.006 per cent carbon and as 16.0 to 16.6 for steels containing 0.17 to 0.59 per cent carbon.

Perhaps the most comprehensive measurements of expansion of iron-carbon alloys were conducted by Driesen.[28] The alloys contained from 0.06 to 3.8 per cent carbon and were low in manganese, silicon, sulphur, and phosphorus. The measurements of expansion seem to have been conducted with care, consequently the expansion coefficients should be fairly reliable although they are certainly not good to the eight decimal places reported. However, they are reproduced here as given in the original.

In Table 97 are given coefficients computed for intervals ranging from 20°C. to a series of temperatures up to 1000°C. The coefficients of Table 98 are for 100°C. intervals. The changes of sign in these columns are a consequence of transformation. Inspection of Driesen's data leads to the conclusion that the coefficient of expansion of ferritic carbon steels declines somewhat in value with increased carbon content and that the coefficient of austenitic carbon steels increases with increased carbon content. The coefficient of expansion of cast iron, at least in the ferrite range, seems to be smaller than that of steel.

D. OTHER PROPERTIES

There remain for discussion in this chapter several properties of iron-carbon alloys which are related vaguely if at all. These are vapor pressure, emissivity, and thermal conductivity. Emissivity is of interest in the measurement of temperature by optical methods, and thermal conductivity is a necessary datum in heat-transfer computation.

216. Vapor Pressure.—Experimental data on the vapor pressure of iron are meager. Jones, Langmuir, and Mackay[214] measured the rate of evaporation of filaments of iron at temperatures from 997 to 1307°C. (1825 to 2385°F.) and computed vapor pressure from a theoretical relation among rate of evaporation, vapor pressure, and temperature. Their vapor pressures (shown in Fig. 200) extrapolate to a normal boiling point of 3200°C. (5790°F.). This value is rounded to 3000°C. (5430°F.) in "The Metal—Iron."[794]

There is apparently no information on iron alloys, but there is little reason to believe that iron-carbon alloys should have vapor pressure much lower than that of iron. Vaporization of iron under ordinary steel-melting conditions should be negligible. In vacuum melting, there is some volatilization, and pressures less than about 0.03 mm. of mercury over molten iron should not be obtainable.

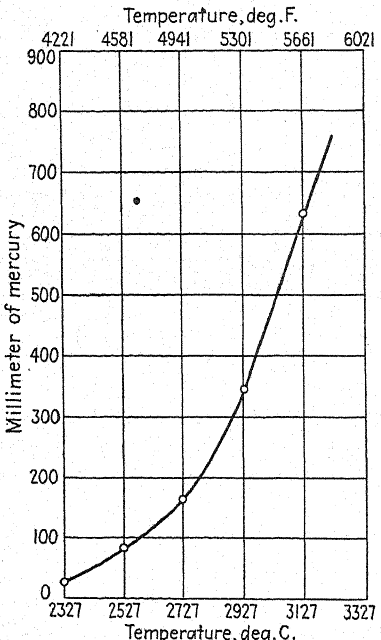

FIG. 200.—Vapor pressure of iron. (*Jones, Langmuir, and Mackay.*[214])

217. Emissivity.—A property of importance in the pyrometry of steel is the emissivity. Inasmuch as unprotected iron and carbon steel soon become covered with rust in an ordinary atmosphere, the room-temperature optical properties are of little interest.

Mellor[741] summarized the results of many workers on the reflectivity of an iron surface. For long wave lengths, the reflectivity ranges from 65 per cent at 1 μ to 96 per cent at 14 μ. For short wave lengths, the reflectivity ranges from 22 per cent at 1880 Å. to 63 per cent at 8700 Å. Chickaschige and Uno[359] showed that the reflecting power does not vary appreciably with carbon content, or with heat treatment.

At elevated temperatures a smooth, clean, bright, metallic iron or steel surface has relatively low emissivity; large correction factors are required in optical and radiation pyrometry.

According to Burgess and Foote,[32] oxidized iron emits 80 to 85 per cent as much radiation as a black body from 700 to 1100°C. (1290 to 2010°F.).

Foote and Fairchild[56] stated that a surface of solid iron oxide in the open gives correct readings with an optical pyrometer at 700°C. (1290°F.), gives readings 5°C. (10°F.) low at 1050°C. (1920°F.), and 10°C. (20°F.) low at 1200°C. (2190°F.). Bright

molten iron, however, gives readings of 1100°C. (2010°F.) when the true temperature is 1185°C. (2165°F.), and 1750°C. (3180°F.) when it is actually 1935°C. (3515°F.). With a radiation pyrometer, the readings are 400°C. (720°F.) low at a true temperature of 1200°C. (2190°F.), and 550°C. (990°F.) low at 1750°C. (3180°F.).

If the mirror surface of a molten metal becomes covered with an oxide film, the emissivity, and hence the apparent temperature, generally increases. Wensel and Roeser[283] found a change to occur at about 1375°C. (2505°F.) on the surface of molten cast iron in the open; emissivity is higher below 1375°C. (2505°F.) and lower above this temperature, consequently, a correction of about 40°C. (70°F.) is required below 1375°C. (2505°F.). Larger corrections are required above 1375°C. (2505°F.); *viz.*, 110°C. at 1400°C. (200°F. at 2550°F.) and 140°C. at 1600°C. (250°F. at 2910°F.). They deduced, therefore, that there is an invisible oxide film at the lower temperature, which dissolves in the metal above 1375°C. (2505°F.). Hase[375] came to a similar conclusion.

218. Thermal Conductivity.—Metals and alloys are used frequently under conditions in which the flow of heat through them is important. Often, however, the rate of heat transfer does not depend upon the properties of the metal, but upon the other materials which are in the path of flow. When the path through the metal is relatively long the conductivity of the metal may have an important influence on the heat transfer. This question has been discussed by Van Dusen.[497]

Judged by the lack of agreement among various determinations, it must be concluded that the thermal conductivity of high-purity iron is not known with any great accuracy. The frequently quoted value of Honda and Simidu[44] of 0.134 cal. per sec. per sq. cm. (°C. per cm.) is certainly too low. Masumoto[218] found 0.174 cal. per sec. per sq. cm. (°C. per cm.) to be representative of a number of values for "fairly pure" iron. [The value chosen in "The Metal—Iron" is 0.19 cal. per sec. per sq. cm. (°C. per cm.) at 0°C.] The effect on thermal conductivity (λ) of additions to iron was given by Masumoto in the form

$$\frac{1}{\lambda} = 5.744 + 2.432\mathrm{C} + 5.087\mathrm{Si} + 2.461\mathrm{Mn}$$

where C, Si, and Mn represent the percentages of the corresponding elements.

There is available in the literature a large number of data on iron-carbon alloys, on, in general, isolated samples. Comparisons among these data in an effort to establish trends are of little value because of the large discrepancies. Donaldson,(604) over a period of years, made careful thermal-conductivity measurements with apparatus capable of variation of no more than about 2 per cent. His conclusions are as follows:

1. The thermal conductivity of wrought iron (trace carbon, 0.92 per cent silicon, 0.20 per cent manganese, 0.014 per cent sulphur, 0.007 per cent phosphorus) is approximately 0.175 cal. per sec. per sq. cm. (°C. per cm.) at 100°C. (210°F.) and decreases with increasing temperature.
2. The thermal conductivity of steel decreases with increasing carbon content, a value of 0.160 cal. per sec. per sq. cm. (°C. per cm.) being obtained for a 0.10 per cent carbon steel and 0.117 cal. per sec. per sq. cm. (°C. per cm.) for a 1.09 per cent carbon steel.
3. The falling off in thermal conductivity with increasing carbon content results from the influence which the carbon has on the structure in producing relative amounts of ferrite, pearlite, and cementite.
4. The data obtained indicate that the thermal conductivity of ferrite is higher than the value 0.174 cal. per sec. per sq. cm. (°C. per cm.) hitherto adopted; and that pearlite has a thermal conductivity of approximately 0.124 cal. per sec. per sq. cm. (°C. per cm.).
5. Black-heart malleable iron, with a structure of ferrite, temper carbon, and a small proportion of pearlite, has a thermal conductivity of 0.150 cal. per sec. per sq. cm. (°C. per cm.) at 100°C. and white-heart malleable iron, with a structure of pearlite, temper carbon, and a small proportion of ferrite, a thermal conductivity of 0.115 cal. per sec. per sq. cm. (°C. per cm.) at a similar temperature.
6. The influence of silicon is to lower the thermal conductivity of all iron alloys. Its influence is most marked in cast iron, where the thermal conductivity decreases rapidly with an increase in silicon content. With over 2 per cent of silicon the falling off in conductivity is reduced considerably, owing to the formation of a ferrite structure.
7. The influence of phosphorus is to produce a slight decrease in the thermal conductivity of gray cast iron.

Shelton and Swanger(658) and Shelton(771) reported the results given in Fig. 201. Except for quenched plain carbon steel (which became tempered during the test), the thermal con-

ductivity decreased linearly with increased temperature. The decrease for cast iron was less rapid than for relatively pure

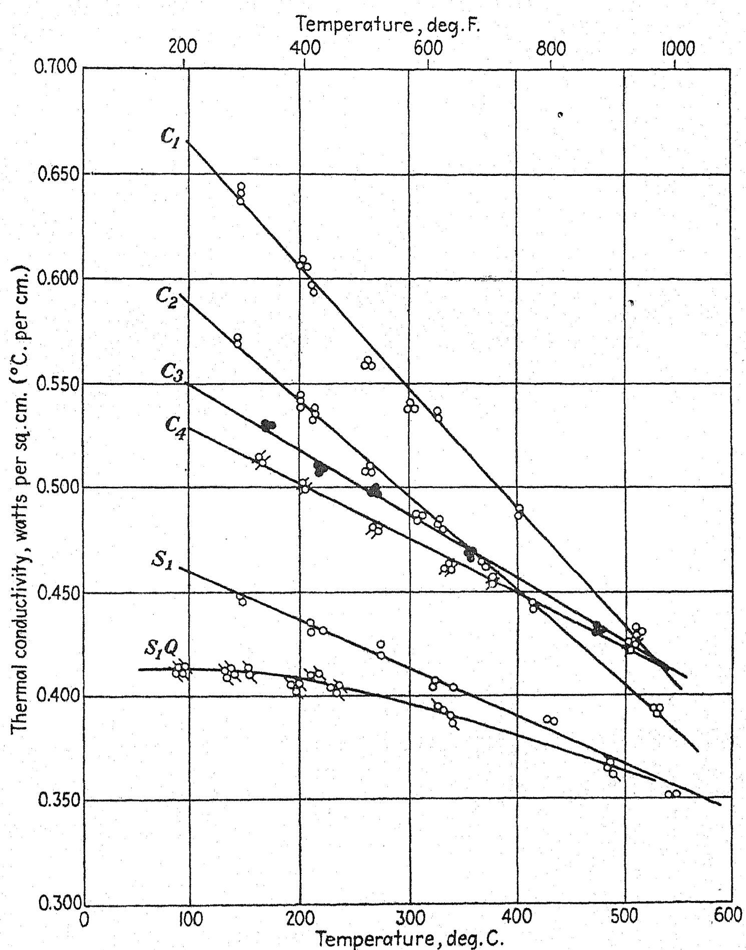

FIG. 201.—Thermal conductivity from 100 to 550°C. (210 to 1020°F.) of: (C_1) open-hearth iron, approximately 99.9 per cent iron; (C_2) wrought iron, approximately 99.5 per cent iron; (C_3 and C_4) cast irons, approximately 4.0 per cent carbon and 1.5 per cent silicon; (S_1) 0.83 per cent carbon steel; and (S_1Q) 0.83 per cent carbon steel as quenched. 1 watt per sq. cm. (°C. per cm.) = 0.239 cal. per sec. per sq. cm. (°C. per cm.) or 57.9 Btu. per hr. per sq. ft. (°F. per ft.). (*Shelton and Swanger*(658) *and Shelton.*(771))

iron, so that at 550°C. (1020°F.) the cast iron was as good a conductor as ingot iron. High-chromium steels, including 18 per

cent chromium, 8 per cent nickel steel, had low conductivity but it increased rapidly with temperature.

As a practical measure of comparative magnitudes, with copper as a standard [for which thermal conductivity at room temperature is about 4 watts per sq. cm. (°C. per cm.) or about 0.96 cal. per sec. per sq. cm. (°C. per cm.)], the conductivity of substantially pure iron is about 1/6 that of copper, of high-carbon steel about 1/9, and of 18 per cent chromium, 8 per cent nickel steel about 1/25.

As Fig. 201 shows, martensitic carbon steel has lower thermal conductivity than the same material in the tempered condition.

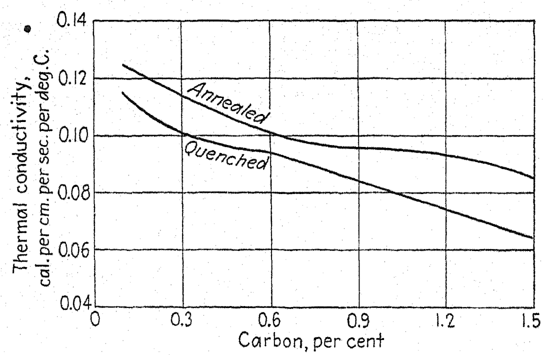

FIG. 202.—Thermal conductivity at 20°C. (70°F.) of iron-carbon alloys containing about 0.5 per cent manganese and 0.2 per cent silicon. (*Naeser.*[274])

Curve S_1Q shows how it tempered during the thermal-conductivity tests. The thermal conductivity of iron, steel, and cast iron decreases with increasing temperature; that of the high-chromium steels, including 18 per cent chromium, 8 per cent nickel steel, increases with increasing temperature. Data obtained by Naeser[274] are given in Figs. 202 and 203.

The thermal conductivity of cast iron is less than that of relatively pure iron at room temperature, but decreases less rapidly with increasing temperature; at 550°C. (1020°F.) cast iron and ingot iron have about the same conductivity. Shelton and Swanger's curves for cast irons of about 4 per cent total carbon, 0.6 per cent combined carbon, and 1.4 per cent silicon are always above and approximately parallel to those for eutectoid steel. Most data on cast irons of lower carbon and higher silicon contents lie in about the same range as those on carbon steels.

According to Donaldson,[364] the thermal conductivity of plain carbon steels (including cast steels) varies from 0.12 to 0.14 cal. per sec. per sq. cm. (°C. per cm.); it decreases with increased carbon content and with increased temperature. The conductivity of gray cast iron varies from 0.11 to 0.137. Free ferrite has a higher thermal conductivity than pearlite, as may be inferred from his values of 0.174 cal. per sec. per sq. cm. (°C. per cm.) for ferrite and of 0.017 cal. per sec. per sq. cm. (°C. per cm.) for cementite.[517] Masumoto[218] found that chilled

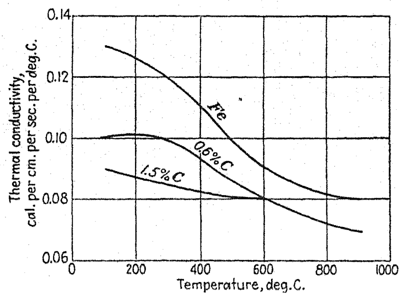

FIG. 203.—Thermal conductivity of iron and two carbon steels at different temperatures. (*Naeser*.[274])

cast iron increased in thermal conductivity during malleablization by 4 per cent for each 1 per cent combined carbon in the chilled iron.

Other data on thermal conductivity were given by Söhnchen,[775] who found that cementite reduces the conductivity of cast iron more than does graphite, by Donaldson,[688] and by Hattori,[710] who discussed tool steel.

E. AUTHOR'S SUMMARY

1. The elastic constants of iron and iron-carbon alloys—modulus of elasticity, Poisson's ratio, torsional modulus, and bulk modulus—as determined at room temperature—are but little affected by composition or structure. The modulus of elasticity is between 29 and 30 million lb. per sq. in.; data are given which indicate that for high-purity iron the modulus is about 30 million lb. per sq. in. and slightly lower for steels containing up to 1 per cent carbon. Poisson's ratio for carbon steels is about 0.3; the variations in this value as reported in the literature cannot be associated with changes in composition and

structure. The modulus of rigidity is approximately 11 million lb. per sq. in. There are no data for the bulk modulus of carbon steels, but no appreciable change with carbon content is to be expected.

2. The density of iron-carbon alloys is sensitive to changes in composition, structure, and temperature, and to phase changes. The density decreases with increasing carbon (cementite), with quenching, and with cold working.

3. In general, iron-carbon alloys expand on melting and contract on solidification. During solidification, the evolution of gas and the formation of graphite may lead to an apparent increase in volume.

4. The thermal expansion of steel is in general somewhat less than that of high-purity iron; that of cast iron may be considerably less. The coefficient of expansion of ferritic carbon steels becomes somewhat lower as the carbon content increases. The coefficient of austenitic carbon steels increases with increased carbon.

5. The vapor pressure of iron increases rapidly above 2500°C. (4530°F.); the boiling point is about 3000°C. (5430°F.).

6. Bare molten iron or steel has a relatively low emissivity; this necessitates a large correction factor in optical and radiation pyrometry. When the metal is covered with a layer of iron oxide, however, the emissivity is fairly high; different slags vary in their emissivity.

7. The thermal conductivity of iron is about one-sixth that of copper. The thermal conductivity of high-purity iron is considerably higher than that of steel. Cast iron has a somewhat higher conductivity than high-carbon steel since cementite and pearlite lower the conductivity more than does graphite. Carbon in solution has the greatest effect in lowering the thermal conductivity; quenched steel, therefore, has lower conductivity than annealed steel.

CHAPTER XVI

ELECTRIC AND MAGNETIC PROPERTIES OF COMMERCIAL IRON-CARBON ALLOYS

Electric Properties—Magnetic Properties—High-carbon Alloys: Permanent Magnets—Author's Summary

Metals and alloys are conductors of electricity but they differ greatly in their capacity to conduct or, conversely, to resist the passage of electric current. For example, the resistivity of high-purity iron (9.8 microhm-cm. at 20°C.) is much higher than the resistivity of copper (1.72 microhm-cm. at 20°C.). Moreover, the resistivity of a metal is usually increased to a marked degree by alloying or by the presence of impurities. The effect of impurities, especially carbon, on the electric properties of iron has been summarized by Cleaves and Thompson.[(794)] Some of their data, enough to serve as a basis of comparison with the electric properties of commercial iron-carbon alloys, are given below.

Of the common chemical elements only three—iron, cobalt, and nickel—possess the property of ferromagnetism. That the most common, and the cheapest, of all the metals (iron) possesses this property is fortunate because, as Cleaves and Thompson stated, "it is largely by the magnetization of iron that the commercial conversion of mechanical to electric power, or of electric to mechanical power, is effected." Although carbon steels have been supplanted for many purposes by other ferrous alloys with better magnetic properties, they have the advantages of cheapness and almost universal availability and are, therefore, still widely used.

The effect of impurities, especially carbon, on the magnetic properties of iron has also been summarized adequately by Cleaves and Thompson. To serve as a foundation for a discussion of the magnetic properties of carbon steels, some of Cleaves and Thompson's data for high-purity iron and for iron containing small amounts of carbon are also summarized in the present chapter.

Methods for the determination of electric and magnetic properties have been used frequently by investigators of the constitution of ferrous alloy systems to determine transformation and transition points. This has been discussed in Volume I of this monograph (pages 29 to 33).

A. ELECTRIC PROPERTIES

The electric property of ferrous materials which is of greatest commercial interest is conductance, or its reciprocal, resistance. The latter, most frequently reported as *specific resistance* or *resistivity*, is the resistance of a standard section, usually a block 1 sq. cm. in cross-sectional area and 1 cm. long; its dimensions are microhms per centimeter cube, or microhm-centimeters. Since the electric resistivity of a ferrous alloy changes with temperature, the temperature must be stated. The change in resistance for a change of 1° (always °C. for scientific work) is called the *temperature coefficient of electric resistance.*

The thermoelectric properties of iron-carbon alloys are of relatively small importance. For iron, these have been reviewed in detail by Cleaves and Thompson, some of whose data on the effect of carbon are summarized in the following sections.

219. Resistivity of High-purity Iron.—In general, the resistivity of metals and alloys increases with temperature; for many pure metals the resistivity is roughly proportional to the absolute temperature. In the absence of transformations it is possible to express the variations of resistivity with temperature by a power series of the form:

$$R_t = R_0(1 + \alpha t + \beta t^2)$$

where R_t and R_0 are resistivities at the temperatures t and 0°C. respectively and α and β are constants, α being much larger than β. This form, however, finds little use. More commonly, the temperature coefficient of resistance is defined by the equation:

$$\text{Temperature coefficient } \alpha = \frac{R_t - R_{t'}}{R_0(t - t')}$$

where R_t, $R_{t'}$, and R_0 are the resistance values at temperatures t, t', and 0°C. The unit of the temperature coefficient is ohms per ohm per degree centigrade, or simply per degree.

The values selected by Cleaves and Thompson[794] for high-purity iron for resistivity at 20°C. (R_{20}), and the temperature coefficient for the range 0° to 100°C. ($\alpha_{0 \text{ to } 100}$) are:

$$R_{20} = 9.8 \text{ microhm-cm.}$$

and

$$\alpha_{0 \text{ to } 100} = 0.0065 \text{ per °C.}$$

This value for the temperature coefficient is abnormally high, being exceeded only by nickel and cobalt. For most pure metals, it is about 0.004 per deg.

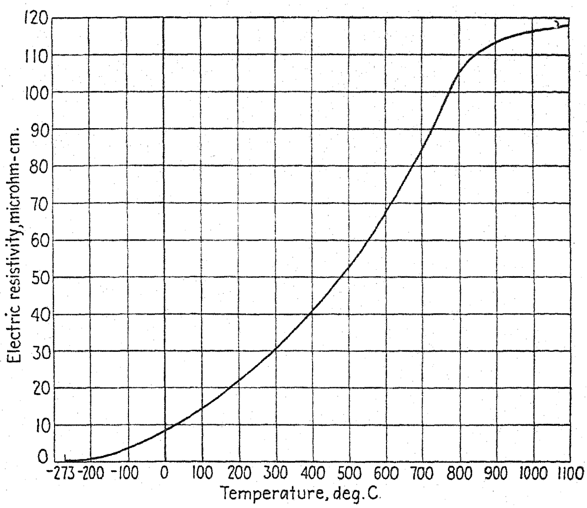

FIG. 204.—Variation of electric resistivity of high-purity iron with temperature. (*Cleaves and Thompson's*[794] *selected values.*)

220. Effect of Temperature on Resistivity of Iron and Iron-carbon Alloys.—The effect of elevated and subnormal temperatures on the resistivity of high-purity iron is shown in Fig. 204 from Cleaves and Thompson.[794] Iron does not become superconducting at temperatures above absolute zero. The resistance approaches zero at absolute zero; its failure to attain zero value, shown in Fig. 204, may be caused by impurity.

Resistivity determinations of iron-carbon alloys at elevated temperatures and at the melting point have been made by Bornemann and Wagenmann.[27] Their values are given in Table 99. The temperature coefficients are given in Table 100.

221. Influence of Carbon and Other Elements on the Resistivity of Iron.—It is well known that the addition of other elements increases the resistivity of iron. Table 101 gives the values of Norbury,(62) as quoted by Cleaves and Thompson,(794) and from

TABLE 99.—EFFECT OF TEMPERATURE ON RESISTIVITY OF IRON-CARBON ALLOYS*

Carbon, per cent	Melting point, °C.	Resistivity, microhm-cm., at melting point	Resistivity, microhm-cm., at temperature of (°C.)							
			1200	1300	1400	1450	1500	1550	1600	1650
0.00	1505	131.1	.....	.....	.....	.....	.....	133.3	135.7	138.1
0.22	1495	136.4	.....	.....	.....	.....	136.6	138.7	140.8	142.9
1.19	1416	149.1	.....	.....	.....	150.1	151.5	152.9	154.3	155.7
3.80	1190	148.0	148.2	150.3	152.6	153.7	154.8	155.9	157.0	

* Bornemann and Wagenmann.(27)

Abegg's handbook.(427) Benedicks'(7) rule that, for all substances which form solid solutions with iron, the change in resistivity is the same for the addition of 1 atomic per cent, is, from the values given in Table 101, seen to be only approximate, even if the ferromagnetic elements nickel and cobalt are omitted.

TABLE 100.—MEAN TEMPERATURE COEFFICIENT OF RESISTANCE (FROM MELTING POINT TO 1650°C.) OF IRON-CARBON ALLOYS*

Carbon, per cent	Melting point, °C.	Mean temperature coefficient α
0.00	1505	0.000366
0.22	1495	0.000308
1.19	1416	0.000188
3.80	1190	0.000148†

* Bornemann and Wagenmann.(27)
† Melting point to 1600°C.

The application of such figures to the computation of the resistivity of steels may provide a reasonably satisfactory approximation, but only that; for, aside from the uncertainty in the individual values, it is known that the effect of additions is not strictly proportional to the amount and, further, may

depend upon other elements present. As these results refer to 20°C. (70°F.) the resistivity of iron at 20°C. (9.8) should be used.

TABLE 101.—EFFECT OF ADDED ELEMENTS ON THE RESISTIVITY OF IRON

Element	Increase in resistivity, microhm-cm.			
	Per 1 weight per cent added		Per 1 atomic per cent added	
	Norbury*	Abegg[427]	Norbury*	Abegg[427]
Carbon	34.0	34.0	7.6	7.6
Manganese	5.0	5.0 to 10.5	4.9	4.9 to 10.3
Silicon	13.5	13.0 to 15.8	6.9	6.5 to 8.0
Sulphur	....	12.0	...	6.9
Phosphorus	11.0	11.0	6.1	6.1
Nickel	1.5	1.55 to 4.55	1.5	1.7 to 4.7
Chromium	5.4	2.5 to 5.4	5.0	2.3 to 5.0
Vanadium	5.0	6.7	4.6	6.1
Tungsten	1.5	2.0 to 3.6	4.9	6.5 to 11.8
Molybdenum	3.4	3.4	5.8	5.8
Cobalt	1.0	1.0 to 3.0	1.0	1.1 to 3.2
Titanium	....	1.0	...	0.9
Copper	4.0	3.0 to 4.0	4.5	3.4 to 4.6
Nitrogen	14.6	14.6	3.8	3.8
Aluminum	12.0	11.1 to 14.4	5.8	6.0 to 7.7

* Selection of values by Norbury[62] as most probable. Quoted by Cleaves and Thompson.[794]

According to Gumlich[47] the resistivity at 20°C. (70°F.) of steels containing up to 0.9 per cent carbon is given by

$$\rho_{20} = 10.5 + 3p + 2p^2,$$

where p is weight percentage carbon. Yensen[144] distinguished between carbon all in solution and carbon as free cementite by means of two formulas:

$\rho_{20} = 9.6 + 82.5p$ (p less than 0.02 per cent carbon)
$\rho_{20} = 11.25 + 4.5(p - 0.02)$ (p greater than 0.02 per cent carbon)

The effect of form of cementite on the resistivity of iron-carbon alloys was investigated by Bardenheuer and Schmidt.[247] Their results for pearlitic cementite and massive cementite are

given in Fig. 205; as may be expected, pearlite is the more effective in increasing the resistivity.

Resistivity values for several cast irons reported by Partridge[276] (Table 107, page 610) range from about 45 to about

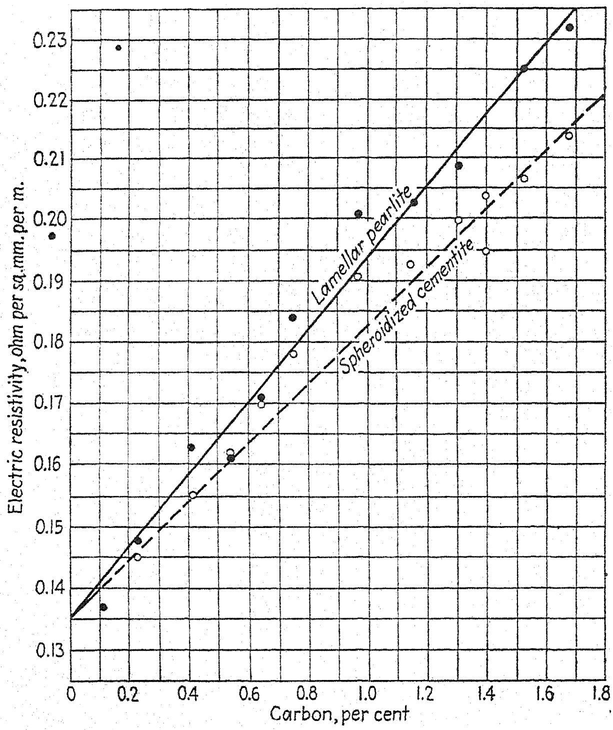

FIG. 205.—Effect of carbon on the electric resistivity of carbon steels with the carbon as lamellar pearlite or spheroidized cementite. (*Bardenheuer and Schmidt.*[247])

60 microhm-cm. The range for malleable iron, given by the Symposium on Malleable Iron Castings,[429] is shown by the following data:

Specimen number	Number of tests	Resistivity, microhm-cm.		
		Maximum	Minimum	Average
1	37	37	28	30.5
2	9	35.02	31.47	33.3

222. Effect of Thermal Treatment on the Electric Resistivity of Carbon Steel.—The effect of quenching on the electric resistiv-

TABLE 102.—COMPOSITION OF SPECIMENS USED BY CAMPBELL(33) IN HIS RESISTIVITY MEASUREMENTS

Element, per cent				
C	Mn	P	S	Si
0.04	0.10	0.007	0.029	
0.30	0.204	0.012	0.013	0.033
0.35	0.08	0.009	0.024	0.18
0.41	0.08	0.012	0.016	0.19
0.57	0.11	0.010	0.020	0.17
0.94	0.27	0.018	0.020	
1.04	0.214	0.019	0.023	0.191
2.71	0.08	0.012	0.014	0.09

ity of commercial iron-carbon alloys was investigated by Gumlich.(47) Some of his results are reproduced in Fig. 206. (The temperature coefficients are for the range 20 to 100°C.) Results

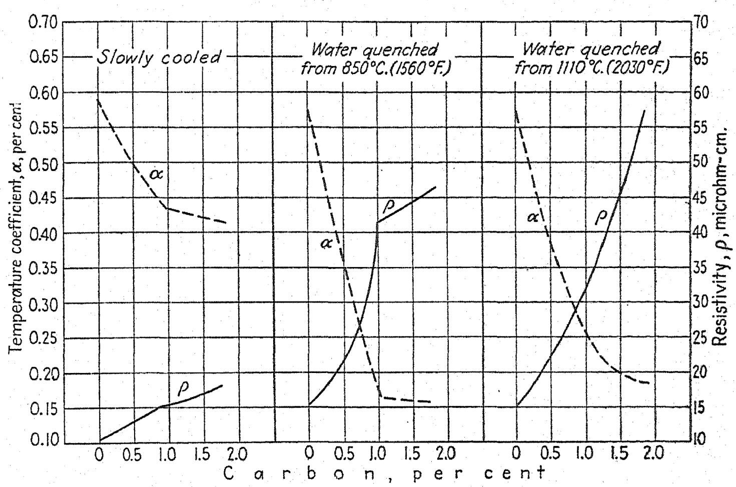

FIG. 206.—Electric resistivity (ρ) and temperature coefficient (α) of iron-carbon alloys after different treatments. (*Gumlich.*(47))

of Campbell(33) on alloys of composition given in Table 102 are collected in Table 103.

TABLE 103.—RESISTIVITY (MICROHM-CM.) OF IRON-CARBON ALLOYS AFTER DIFFERENT QUENCHING AND TEMPERING TREATMENTS*

Carbon, per cent	Quenched from		Steel quenched from 890°C. (1635°F.) reheated to							
	1105°C. (2020°F.)	890°C. (1635°F.)	105°C. (220°F.)	195°C. (385°F.)	295°C. (565°F.)	400°C. (750°F.)	490°C. (915°F.)	600°C. (1110°F.)	700°C. (1290°F.)	800°C. (1470°F.)
0.04	.....	11.12	10.94	10.81	10.83	10.78	10.75	10.69	10.78	10.70
0.30	16.48	16.13	15.56	14.96	13.86	13.55	13.04	12.84	12.64	13.20
0.35	18.45	17.59	16.67	16.05	15.11	14.86	14.37	14.25	14.27	14.51
0.41	20.05	18.91	17.79	16.94	15.69	15.13	14.82	14.51	14.47	14.98
0.57	23.60	22.31	19.92	17.80	16.24	15.60	15.02	14.71	14.67	15.11
0.94	.....	25.95	27.97	22.40	19.42	18.37	17.60	16.65	16.19	17.23
1.04	48.43	41.92	33.02	25.73	21.82	20.62	19.92	19.60	19.03	19.08
2.71	58.73	52.59	42.65	32.49	27.41	26.17	24.94	24.21	22.57	20.82

* Campbell.(33)

Indication that the resistivity of quenched steels decreases with the passage of time was found by Benedicks and associates.[174] Their data are given in Table 104.

TABLE 104.—EFFECT OF TIME ON THE RESISTIVITY OF STEELS WATER QUENCHED FROM 800°C. (1470°F.)*

Mark	Carbon, per cent	Resistivity, microhm-cm., at 18°C. (65°F.)		
		1900	1919	1926
1	0.08	10.90	10.6	10.79
2	0.45	28.99	27.9	28.33
3	0.55	34.36	32.9	32.90
4	0.90	36.94	32.2	32.16
5	1.20	42.13	36.8	36.38
6	1.35	44.35	39.6	39.25

* Benedicks, Bäckström, and Sederholm.[174]

223. Thermoelectric Properties.—The thermoelectric properties of the iron-carbon alloys are of relatively small interest, and there are few data available. The effect of carbon was discussed by Broniewski,[25] Dupuy and Portevin,[34] and Campbell;[42] Campbell discussed also the effect of quenching and tempering. These data are summarized in "The Metal—Iron"[794] and need not be repeated here. The thermal electromotive force increases with increased carbon content, is much greater in the quenched than in the annealed condition and decreases as the quenched specimen is tempered at increasing temperatures, the limit being the temperature of eutectoid transformation.

B. MAGNETIC PROPERTIES

Commercial alloys of iron and carbon, in the earliest days of electric machinery, were the only available core material. Cast iron was much used for frames and yokes; even though it is not a good magnetic material, cast iron is cheap, is easy to cast, and is easy to machine. The magnetic properties of cast steel are much better than those of cast iron, consequently this material has to a great extent supplanted cast iron for such parts as frames and yokes. (It has been demonstrated in recent years, however, that cast iron having relatively good magnetic properties can be produced.) Carbon steel was once the only material

used for construction of transformer cores; it is still used somewhat, because of its cheapness, for laminated poles. The magnetic material of widest application is open-hearth steel containing silicon (electric sheet). One grade of electric sheet contains 1 per cent silicon, or less; *e.g.*, the "armature" grade contains about 0.4 per cent. Carbon, however, is kept as low as possible. From this it may be inferred correctly that iron-carbon alloys are not important *magnetic* materials; they are used, as has been stated, because of cheapness and ease of producing the required shape, usually in the form of cast iron, cast steel, or malleable iron.

It is in the making of permanent magnets that carbon becomes a useful element, although plain carbon steels are seldom used for this purpose in modern times. The best permanent-magnet steels are of the high-alloy type, consequently they are properly discussed in other monographs.

In brief, so far as magnetic properties are concerned of materials for cores of transformers and rotating machinery, the less carbon the better. Permanent-magnet steels require carbon but also substantial amounts of other elements. Therefore, from the practical point of view, there is little to be said about magnetic properties of iron-carbon alloys, except that they are inferior.

224. Ordinary Magnetic Properties of Iron-carbon Alloys.—Of the available data on magnetic properties of iron-carbon alloys, those supplied by Yensen(144) are best able to show the effect of carbon. Yensen's alloys were vacuum melted; his materials were electrolytic iron and graphite. This means that the data of Table 105 are not representative magnetic properties of commercial carbon steels, but they provide probably the best basis of comparison. The manner in which the properties of commercial alloys differ from those given in Table 105 will appear upon examination of specific properties. The reader is referred to Cleaves and Thompson(794) and to Greiner, Marsh, and Stoughton(618) for more complete discussion of the properties of magnetic materials.

225. Saturation Magnetization of Iron-carbon Alloys.—Saturation magnetization of iron-carbon alloys was first reported by Hadfield and Hopkinson(20) and by Gumlich.(23) Decrease of value of saturation magnetization was found in both investiga-

TABLE 105.—MAGNETIC PROPERTIES OF VACUUM-MELTED ALLOYS OF ELECTROLYTIC IRON AND GRAPHITE*

Number	Carbon, per cent†	Maximum permeability	Induction, kilogausses, for						Saturation magnetization, $4\pi I_\infty$, kilogausses §	For B = 9650 gausses		
			H = 0.2	H = 1	H = 4	H = 20	H = 100	H = 400		Hysteresis loss, ergs per cu. cm. per cycle	Residual induction, gausses	Coercive force, oersteds
2C 201	0.008	9000	0.1	9.0	15.6	17.4	19.0	21.6	22.80	1,700	8100	0.58
202	0.016	6150	0.2	5.9	14.3	16.8	18.6	21.4	22.50	1,960	7100	0.68
203	0.021	5400	0.2	4.8	13.4	16.3	18.2	21.1	22.65	2,070	7000	0.67
204	0.061	4670	0.1	3.8	12.2	15.9	18.0	21.0	22.60	2,150	6550	0.76
205	0.097	4800	0.2	4.5	11.8	15.5	17.5	20.0	.....	2,100	6150	0.74
206	0.174	3200	0.2	2.4	9.6	14.3	16.9	20.0	.....	3,300	6650	1.06
207	0.213	3000	0.1	2.3	9.0	14.2	16.6	19.8	.....	3,500	6400	1.11
208	0.205	2900	0.1	2.3	9.0	14.2	16.8	19.8	.....	3,440	6450	1.11
209	0.283	2400	0.1	1.7	7.9	13.4	16.1	19.2	.....	4,500	6750	1.32
210	0.268	2500	0.1	1.7	8.4	14.1	16.9	20.2	.....	4,260	6800	1.27
211*a*	0.235‡	2350	...	0.8	8.6	13.9	16.2	....	.....	5,060	7100	1.58
211*b*	0.151	3640	0.1	3.0	10.5	14.8	17.0	....	.....	3,100	7100	0.95
212*a*	0.339‡	1790	...	0.6	7.0	13.1	14.4	....	.....	6,630	7000	2.16
212*b*	0.227	2700	0.1	2.4	8.9	13.9	16.4	....	.....	3,820	7000	1.09
213*a*	0.485‡	1260	...	0.4	4.9	12.1	13.7	....	.....	9,100	7250	3.01
213*b*	0.387	1640	...	1.3	6.4	13.0	15.5	....	.....	8,000	7350	2.13
214*b*	0.559	1290	...	1.3	5.1	13.7	16.9	....	.....	9,450	7250	2.85
215*a*	0.850‡	800	...	0.3	1.8	11.4	13.3	....	.....	15,400	7400	4.95
215*b*	0.791	815	...	0.3	2.0	11.8	15.3	....	.....	15,000	7600	4.95
216*a*	0.985	830	...	0.2	1.5	10.3	13.7	....	.....	17,300	7700	5.60
216*b*	0.981	750	...	0.2	1.5	10.2	13.5	....	.....	17,900	8100	5.60

In accordance with a note by Yensen,(144) (p. 174 of his article), 3.5 per cent has been deducted from the original values for ring samples of permeability and induction. To the original values of coercive force and hysteresis loss, 5.5 (3.5[1.6]) per cent has been added.

* Yensen.(144)

† Carbon content after annealing in vacuum at 900 to 935°C. (1650 to 1715°F.).

‡ Annealed in nitrogen.

§ These values are probably high (see p. 596).

tions. These data and those of Yensen,[144] Maurer and Schroeter,[321] Stäblein and Schroeter,[282] Cheney,[82] and Esser and Ostermann[697] are summarized in Fig. 207. The points of Yensen are probably high (see Cleaves and Thompson,[794] page 218). The points of Esser and Ostermann may be suspected also. Extrapolation of Gumlich's high points (curve I) leads to the value 21,600 gausses for iron. This is in agreement with the experimental value of Stäblein and Schroeter for electrolytic iron. Gumlich's computed value of about 12,500 gausses for cementite was also confirmed by Stäblein and Schroeter, who found 12,400 gausses. Esser and Ostermann, however, reported 13,200 gausses. The scatter of the points of Cheney suggests that they are not to be given much weight.

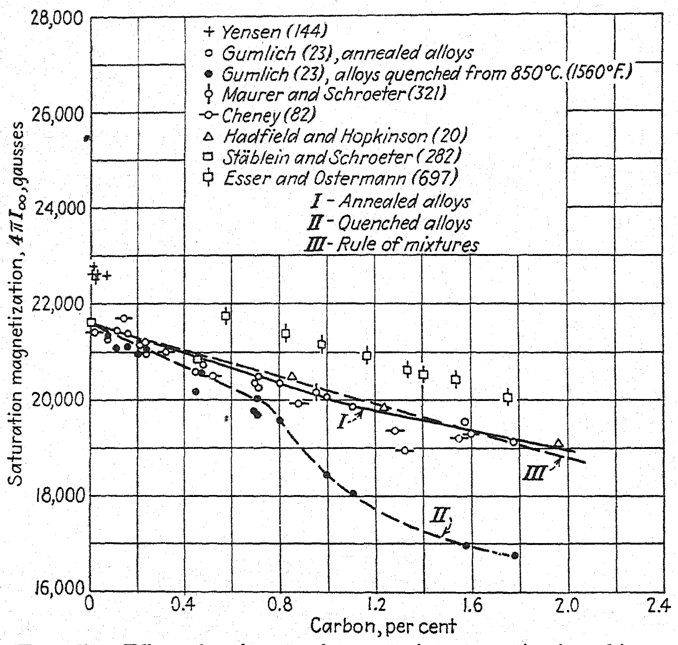

FIG. 207.—Effect of carbon on the saturation magnetization of iron.

The broken line, marked III, of Fig. 207, represents the rule-of-mixtures values. It appears that such values, in the present state of knowledge, are as good as the observed values; consequently, saturation magnetization, $4\pi I_\infty$, of iron-carbon alloys—at least up to 2 per cent carbon—is given by:

$$4\pi I_{\infty} = 21{,}600(1 - 0.15C) + 12{,}400(0.15C) = 21{,}600 - 1{,}380C$$

where C is weight percentage carbon. The factor 0.15 is necessary to convert percentage carbon to percentage cementite.

Curve II, which represents Gumlich's data on alloys quenched from 850°C. (1560°F.), is given merely because such data have been used to estimate the fraction of austenite retained upon quenching; this is shown in Table 106, page 608.

226. Magnetic Induction of Iron-carbon Alloys.—Carbon decreases, for a given field strength, the induction of iron. The effect is especially noticeable in fields of low strength, as is

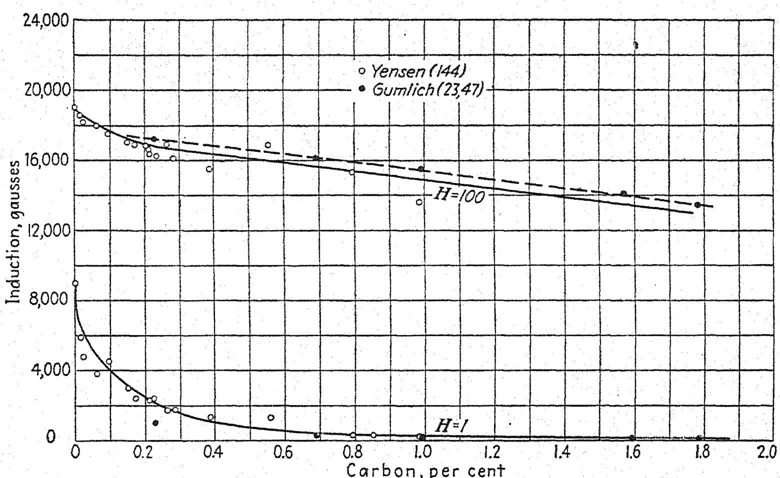

FIG. 208.—Magnetic induction of iron-carbon alloys. Curves indicate trends only.

indicated by the data of Fig. 208 for a field strength of 1 oersted; the effect is less marked at 100 oersteds. Of more interest, however, is the permeability.

227. Maximum Permeability of Iron-carbon Alloys.—The high maximum permeability of purified iron (see Cleaves and Thompson,[794] page 219) is reduced sharply by small additions of carbon; this is demonstrated in Fig. 209. The curve was drawn by giving most weight to the data of Yensen.[144] It is to be seen that as little as 0.05 per cent carbon reduces maximum permeability to about 5000; the decline is less rapid with further increase of carbon content, although the value is less than 1000 for all carbon contents greater than about 0.7 per cent. Gum-

lich's[23] data are in good agreement with those of Yensen, especially if the lower purity of the alloys of the former is considered. Cheney's[82] data agree in trend but exhibit the scatter noticed upon examination of his data on other magnetic properties.

228. Coercive Force and Residual Induction of Annealed Iron-carbon Alloys.—The main measurements of coercive force

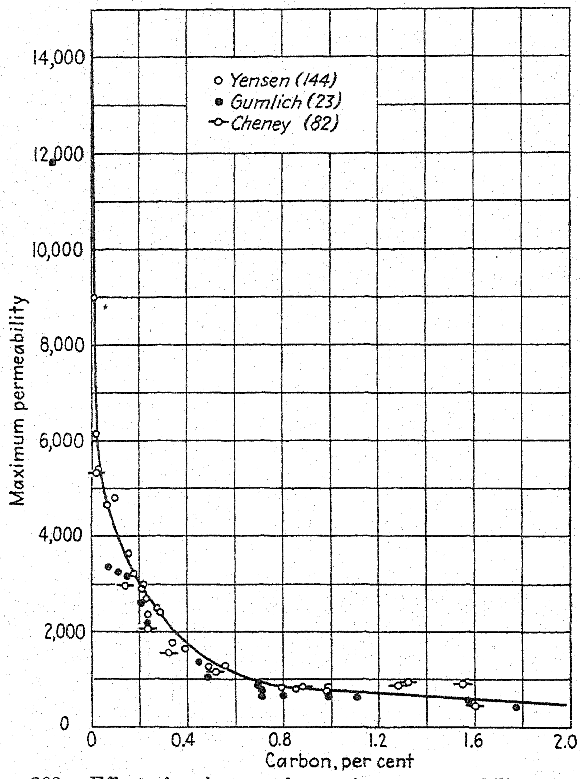

FIG. 209.—Effect of carbon on the maximum permeability of iron.

of iron-carbon alloys in the annealed condition were those of Gumlich[23,47] and of Yensen.[144] Yensen's values for a given carbon content are consistently lower than those of Gumlich, as may be seen by inspection of Fig. 210. The points of Cheney[82] are scattered, and there seems no reason to give credence to the discontinuities indicated. It may be assumed that Yensen's data represent the variation of coercive force with carbon content

of high-purity alloys of iron and carbon. It is to be seen that carbon in solution causes rapid increase of coercive force and that the increase beyond saturation is less rapid. Since Gumlich's alloys were commercial steels—consequently less pure than Yensen's—it is to be concluded that other elements are capable of causing substantial increase of coercive force.

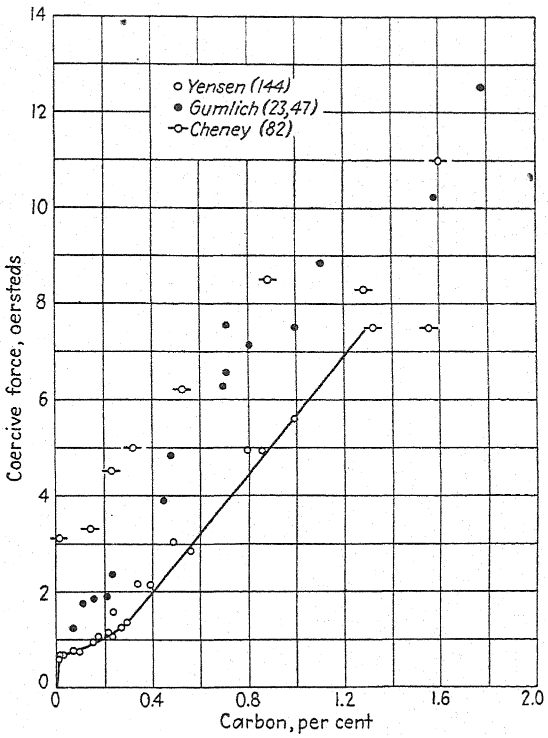

FIG. 210.—Effect of carbon on the coercive force of iron.

Available data do not indicate a clear effect of carbon on residual induction. Yensen's[144] data for an induction of 10 kilogausses indicate residual inductions of 6 to 8 kilogausses. Gumlich's[23,47] data for commercial alloys magnetized to saturation are somewhat higher in value, mostly lying in the range from 10 to 11 kilogausses.

229. Effect of Impurities.—The principal investigations of the effect of impurities on the magnetic properties of iron were conducted by Yensen[581] and by Gumlich.[47] This information

has been summarized adequately by Cleaves and Thompson.[794] In brief, small amounts of elements that enter the lattice of iron interstitially—such as carbon, nitrogen, and oxygen—and generally the substitutional elements that decrease the temperature of A_3 transformation—such as copper, manganese, and nickel—exert an injurious influence, as measured by permeability. Small

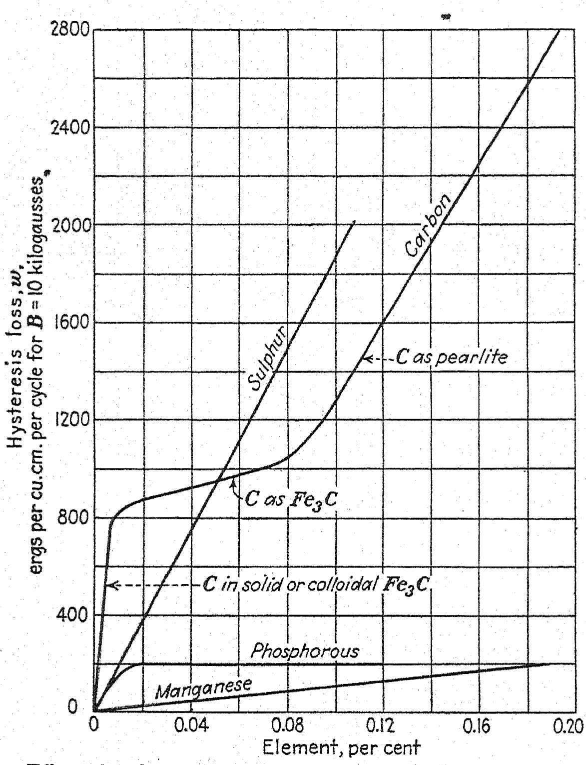

FIG. 211.—Effect of carbon, manganese, sulphur, and phosphorus on the hysteresis loss of iron. (*Yensen.*[581])

amounts of substitutional elements that increase the temperature of A_3 transformation (in general, the "deoxidizers")—such as aluminum, silicon, and vanadium—exert a beneficial influence, as measured by permeability.[501] The effect of the common impurities, as measured by hysteresis loss, as determined by Yensen,[581] is shown in Fig. 211. The linear variation of hysteresis loss with carbon content, beyond 0.1 per cent carbon,

exists to at least 1 per cent carbon; at 1 per cent carbon, the hysteresis loss is about 16,700 ergs per cu. cm. per cycle.

230. Effect of Grain Size and Thickness of Sample.—Information on the effect of grain size on the magnetic properties of iron has been summarized by Cleaves and Thompson.(794) It is sufficient here to state that the weight of the evidence favors the conclusion that large-grained material conduces smaller hysteresis loss and greater permeability than small-grained

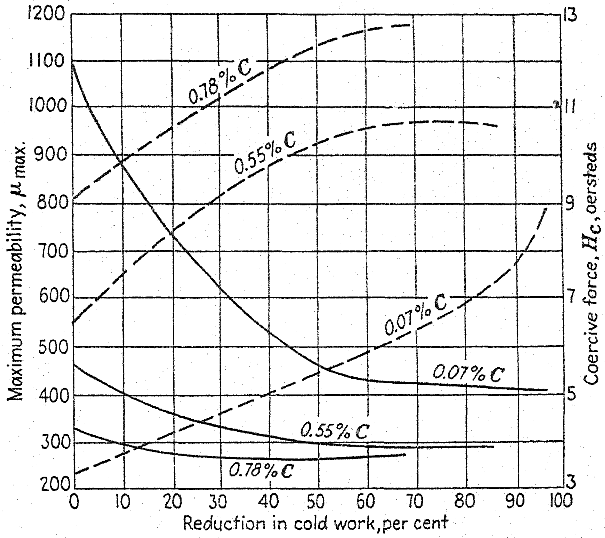

FIG. 212.—Change in maximum permeability (solid lines) and coercive force (dashed lines) of carbon steels with degree of cold work. (*Abegg.*(427))

material. In ordinary materials, however, the grain-size effect may be obscured by the influence of such factors as purity, size, or thermal history of the specimen.

Difference of value of certain magnetic properties as a result of variation of thickness of specimen was found by Cioffi,(513) Spooner,(238) and Ruder.(761) This effect appears to reduce to a matter of grain size or grain number; *e.g.*, Cioffi concluded that if the grains become so large, or if the specimen is so thin, that only a few grains occupy the cross-section perpendicular to the applied field, the flux distribution becomes non-uniform, with the result that the over-all permeability of the specimen falls.

231. Effect of Cold Work on Magnetic Properties.—The permeability of iron-carbon alloys is reduced by cold work; the effect is most marked in low-carbon alloys and with weak fields. The effect of cold work on the maximum permeability

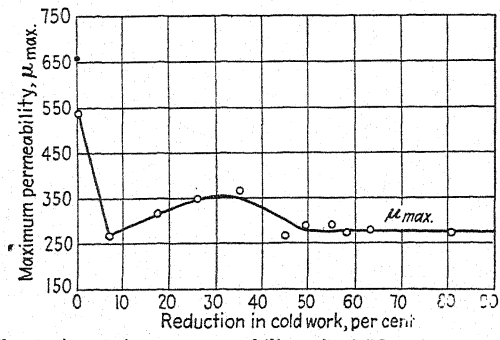

Fig. 213.—Change in maximum permeability of a 0.78 per cent carbon steel with degree of cold work. (*Messkin.*[325])

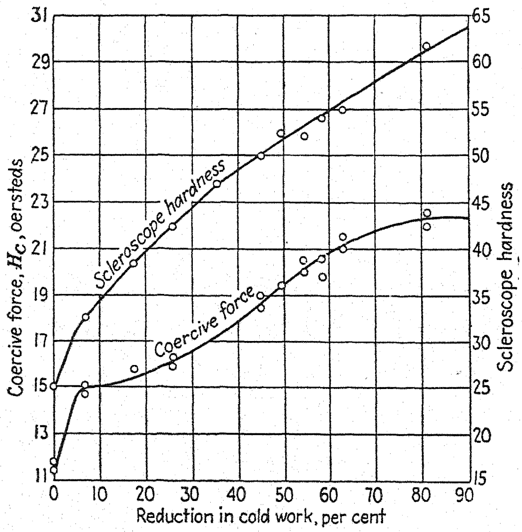

Fig. 214.—Change in coercive force and scleroscope hardness of a 0.78 per cent carbon steel with degree of cold work. (*Messkin.*[325])

of three steels is shown in Fig. 212 from Abegg.[427] It is to be seen also that the decrease of permeability is accompanied by increase of coercive force. Similar results were found by Messkin,[325] as shown by Figs. 213 and 214.

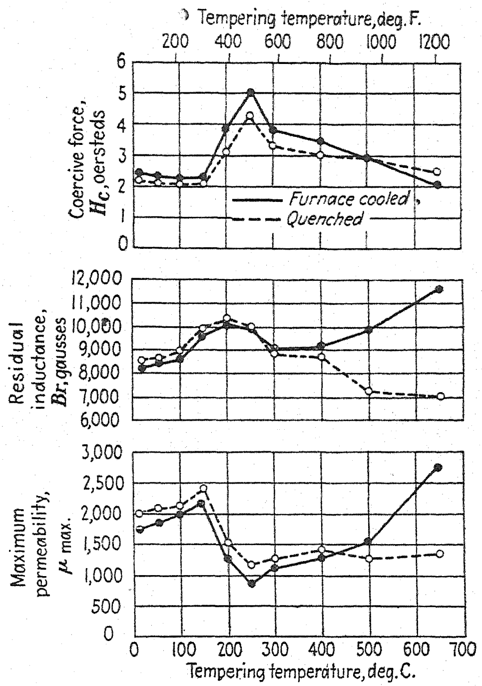

FIG. 215.—Effect of tempering for 1 hr. on the coercive force, residual induction, and maximum permeability of a low-carbon steel water quenched from 680°C. (1250°F.). The effect of the cooling rate after tempering is also indicated. (*Köster*.[313])

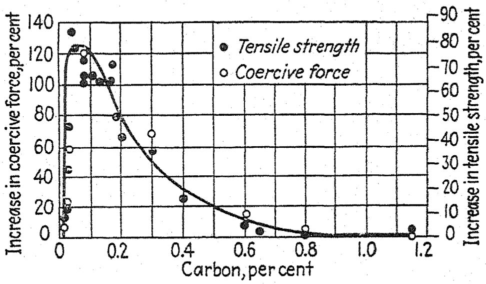

FIG. 216.—Effect of carbon on (1) the increase in coercive force on tempering 1 hr. at 250°C. (480°F.) and (2) the increase in tensile strength after storing 2 weeks at room temperature, of steels previously water quenched from 680°C. (1250°F.). (*Köster*.[313])

232. Effect of Aging.—The aging of low-carbon steel manifests itself by change of magnetic properties as well as change of mechanical properties. This is illustrated by Figs. 215 and 216 from Köster.[313] In Fig. 215 are shown changes upon tempering of coercive force, residual induction, and maximum permeability of a quenched 0.08 per cent carbon steel. To be noted are: (1) the marked increase of coercive force and residual induction in the blue-heat region (about 250°C., 480°F.); and (2) the sharp decrease of maximum permeability in the 150 to 250°C. (300 to 480°F.) range. In Fig. 216 are shown coercive force and tensile strength as a function of carbon content. The steels were quenched from 680°C. (1255°F.), then tempered at 250°C. (480°F.) for 1 hr., or stored at room temperature for 2 weeks (for tensile tests). The high-carbon alloys are evidently insensitive to aging.

C. HIGH-CARBON ALLOYS: PERMANENT MAGNETS

The essentials of a good permanent magnet are high residual induction, high coercive force, and permanence. Magnetic hardness, which may be measured by coercive force, roughly parallels mechanical hardness; heat treatment is therefore necessary to obtain the greatest magnetic hardness. However, the strained condition of untempered martensite and the tendency of carbon to precipitate from freshly tempered martensite make for lack of permanence of the magnet. According to Spooner,[238] the loss of strength by aging of a carbon-steel magnet may amount to as much as 40 per cent. Where permanence is essential, as in electric measuring devices, artificial seasoning treatments—often elaborate—are used.

The usual permanent-magnet material, as was stated before, is alloy steel. The aim of alloy addition is high coercive force without concomitant appreciable loss of residual induction. A possible indication of Mathews'[163] work is that retention of a certain amount of austenite is necessary in order to obtain the highest coercive force.

A frequently used index of value of a permanent-magnet steel is the product of residual induction with coercive force ($B_r \times H_c$). However, the most favorable combination of residual induction and coercive force depends on the dimensions of the magnet. For example, according to Gumlich,[23] if the form is a long,

straight rod, a nearly closed ring, or a horseshoe with a narrow gap, the best results are obtained with the lower carbon (hypoeutectoid) steels; in these forms the more important factor is high residual induction because the demagnetizing effect of the ends is small. If, on the contrary, the form is a short, stout rod or a horseshoe with a broad gap, the higher carbon (hypereutectoid) steels are preferred because of their higher coercive force. It has been accepted generally that a good permanent magnet should have a ratio of length to diameter of 25 to 30.[115]

It is evident that the treatment of a permanent-magnet material is the opposite of that of core material.

In some applications, cost is more important than good magnetic properties. In this event, the expensive-but-good alloy steels may be supplanted by plain high-carbon steel or even cast iron.[238] The properties of annealed iron-carbon alloys have been reviewed briefly already; there remain for discussion properties of alloys in other conditions of heat treatment and the properties of cast iron.

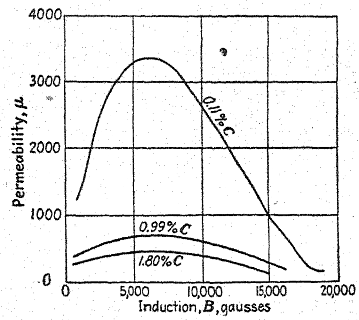

FIG. 217.—Variation of maximum permeability with carbon content for annealed steels. (*Gumlich.*[23])

233. Magnetic Properties of Annealed High-carbon Alloys.—For the purpose of comparison, some information on the properties of annealed alloys is reported here.

Of the two properties used in the customary index of value of a permanent-magnet steel—residual induction and coercive force—only coercive force increases appreciably with increased carbon content. The change of residual induction, if any, is not clear. Maximum permeability of hypereutectoid steels is less than 1000; the induction for maximum permeability, as may be seen from Fig. 217, is about 6500 gausses.

234. Magnetic Properties of Quenched Iron-carbon Alloys.—The effect of quenching from various temperatures on the magnetic properties of a given alloy is shown vividly by Sjovall's[414] hysteresis loops given in Fig. 218 for a ball-bearing steel; these loops, however, correctly portray the behavior of iron-carbon alloys. It is to be seen that no appreciable change is effected

if the quenching temperature is below that of transformation. If the quenching temperature is within the transformation range, coercive force is increased. Quenching from still higher tempera-

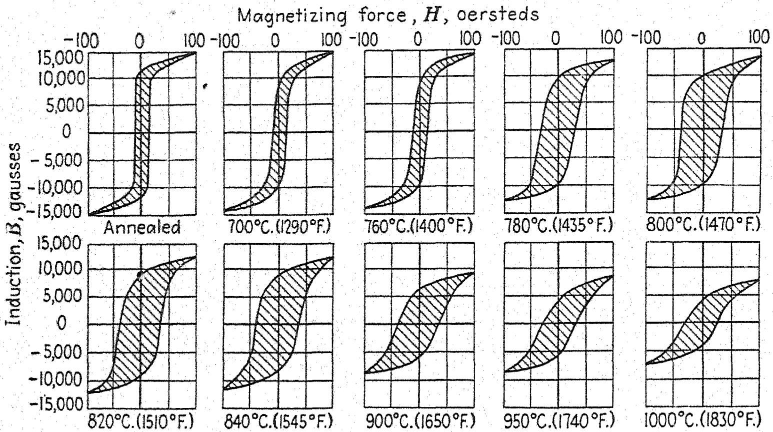

FIG. 218.—Hysteresis curves of ball-bearing steel annealed and then water quenched from various temperatures. (*Sjovall.*[414])

tures results in the retention of austenite, consequently residual and maximum inductions are decreased.

Numerical values of coercive force and residual induction after quenching were given by Gumlich.[23] Some of these data are summarized in Figs. 219 and 220. In Fig. 219 it is to be noted

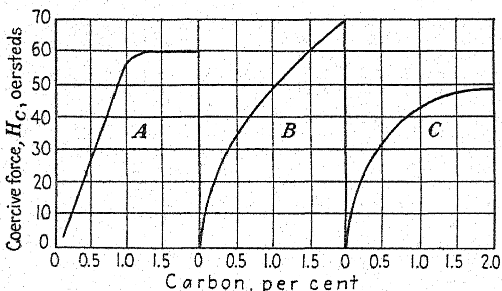

FIG. 219.—Effect of carbon content on coercive force after hardening from *A*—800°C. (1470°F.), *B*—900°C. (1650°F.), and *C*—1000°C. (1830°F.). (*Gumlich.*[23])

that, for a quenching temperature of 800°C. (1470°F.), the coercive force increases rapidly with carbon content to about 1 per cent carbon, then remains constant. The reason for the

difference among the curves for the different quenching temperatures is probably related to the amount of cementite dissolved and the amount of austenite retained, but there is no need here of detailed discussion. In Fig. 220 it is to be noted that the residual induction of alloys quenched from 900°C. (1650°F.) declines rapidly with increased carbon content.

Gumlich's[23] data on saturation magnetization of quenched alloys were given in Fig. 207, page 596. Other data were obtained by Maurer and Schroeter[321] during investigation of retention of austenite by quenching; these are repeated in Table 106. According to their data on 0.95 per cent carbon steel, more austenite was retained by water quenching than by oil quenching. Tamaru and Sekito,[494] however, reported that more austenite was retained by oil quenching of steels containing 0.89 and 1.28 per cent carbon. Thus, information which might be useful to the permanent-magnet maker is inconclusive. It is clear, none the less, that the highest values of coercive force are to be obtained by quenching.

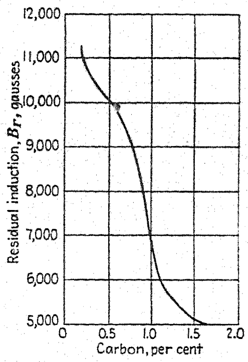

FIG. 220.—Effect of carbon content on residual induction after quenching from 900°C. (1650°F.). (*Gumlich.*[23])

235. Magnetic Properties of Tempered Iron-carbon Alloys.—The effect of tempering on the magnetic properties of quenched iron-carbon alloys was determined by Cheney.[82] According to Fig. 221 the residual induction passed through a minimum for tempering temperatures between 100 and 200°C. (210 and 390°F.). The trend of coercive force, shown by Fig. 222, was downward, although minima appeared for the higher carbon alloys at intermediate tempering temperatures; these seemingly corresponded to small maxima of the maximum-permeability curves (not reproduced here). Results of Dowdell[115] indicate similar trends.

236. Magnetic Properties of Cast and Malleable Irons.—According to Spooner,[238] the maximum permeability of the cast iron ordinarily used in electric machinery is about 500 at an induction of 2 to 5 kilogausses. The saturation induction is about 14 kilogausses. Malleable iron has magnetic properties better than those of cast iron, but not so good as those of cast

steel, for which saturation induction is about 21 kilogausses and maximum permeability is about 1500 at an induction of about 7 kilogausses. Detailed information on the magnetic properties

TABLE 106.—AMOUNT OF RETAINED AUSTENITE IN A CARBON AND IN A CHROMIUM STEEL AFTER WATER AND OIL QUENCHING AS DETERMINED BY MAGNETIC-SATURATION MEASUREMENT*

Quenching medium	Quenching temperature		0.95 per cent carbon steel		0.93 per cent carbon, 2 per cent chromium steel	
	°C.	°F.	$4\pi I_\infty$, gausses	Austenite,† per cent	$4\pi I_\infty$, gausses	Austenite,† per cent
Water	800	1470	17,990	10.5	17,860	1.5
	900	1650	17,180	14.5	16,150	11.0
	1000	1830	17,860	11.5	12,530	31.0
	1100	2010	17,550	13.0	13,320	26.5
	1200	2190	17,790	11.5	13,000	28.5
Oil	800	1470	20,170	0	15,800	13.0
	900	1650	20,180	0	14,750	18.5
	1000	1830	20,150	0	11,780	35.0
	1100	2010	20,120	0	11,840	34.5
	1200	2190	19,550	3	11,920	34.0
Difference between water and oil quench	800	1470	+2,180	....	−2,060	
	900	1650	+3,000	....	−1,400	
	1000	1830	+2,290	....	− 750	
	1100	2010	+2,570	....	−1,480	
	1200	2190	+1,760	....	−1,080	
Annealed specimen	....	....	20,150	....	18,120	

* Maurer and Schroeter.(321)

† $\frac{a-b}{a} \times 100$ = per cent austenite, where a is the saturation value of annealed specimen, and b is the saturation value of tested specimen (quenched as shown).

of cast iron was given by Partridge,(168,276) who concluded that cast iron is not necessarily so bad a magnetic material as was thought. Other conclusions of Partridge seem sufficiently important to quote in part:

The highest magnetic induction and permeability are obtained with cast iron which has been annealed.

The annealing appears to be a very precarious operation. Very slow cooling through the transformation range is required.

Carbon has a very marked influence on the magnetic properties. If present as graphite in the nodular form, high induction is obtained; if in the form of flakes, the specimen possesses low magnetic induction.

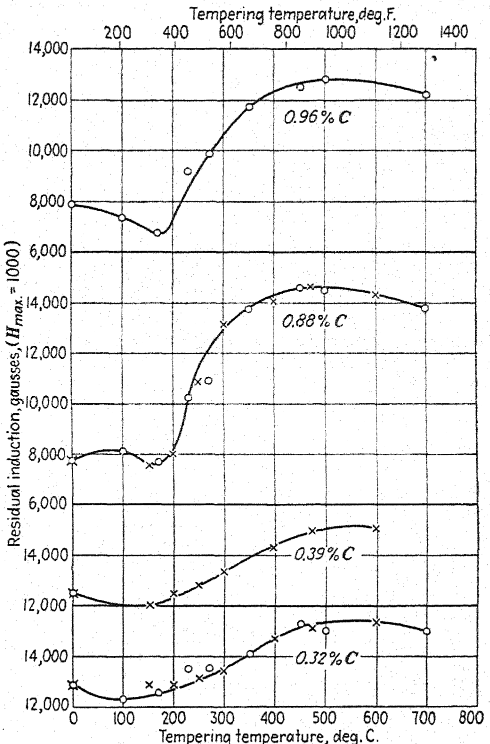

FIG. 221.—Effect of tempering temperature on residual induction of four quenched carbon steels. (*Cheney.*[82])

If the ground mass is ferrite, low hysteresis loss and high permeability appear to be fundamental properties of the material. If the specimen has a pearlite matrix, it will also possess high hysteresis loss and comparatively low permeability.

Graphite does not affect the hysteresis loss, but prevents the attainment of very high magnetic induction. This is due to lack of magnetic continuity of the structure. This lack of continuity causes small demagnetizing forces to act within the specimen. . . .

Some of Partridge's data are reproduced in Table 107 for the purpose of indicating order of magnitudes. Maximum permeability in every instance was increased by annealing.

The foregoing conclusions of Partridge were confirmed in general by more recent work of Söhnchen[774] who found, with

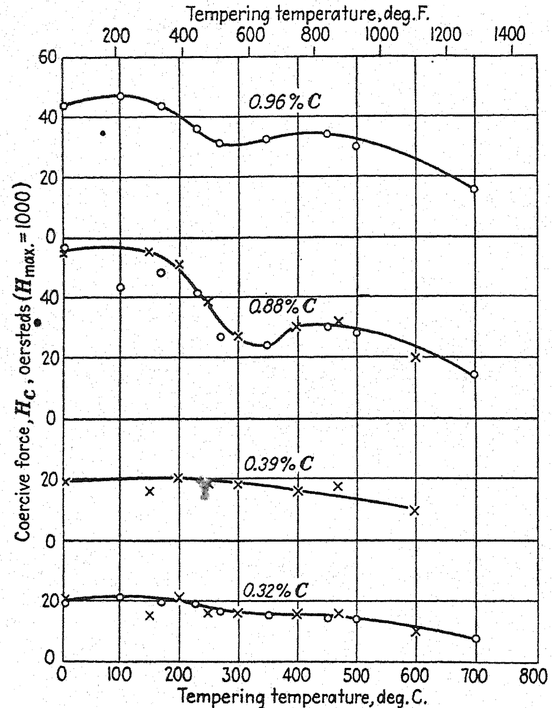

FIG. 222.—Effect of tempering temperature on coercive force of four quenched carbon steels. (*Cheney.*[82])

increased proportion of graphite—which corresponds to annealing—increased maximum permeability and saturation mag-

TABLE 107.—MAGNETIC PROPERTIES OF FOUR IRONS AS CAST*

Composition, per cent				Maximum permeability	Induction, kilogausses, for H = 100	Residual induction, kilogausses	Hysteresis loss		Electric resistivity, microhm-cm.
Total carbon	Graphite	Si	Mn				Ergs per cu. cm. per cycle	Watts per lb.	
2.77	0.89	0.6	0.015	264	9.98	5.03	34,080	21.0	46.8
2.79	0.65	0.81	Trace	237	9.68	4.73	27,680	16.75	50.66
3.11	2.87	1.43	0.024	481	11.48	5.12	26,120	16.44	47.73
2.95	2.6	1.97	0.025	497	11.64	5.52	23,860	14.9	57.15

* Partridge.[276]

netization and decreased coërcive force. In general agreement with the statement on page 599, Söhnchen could find no clear relation between residual induction and graphite content.

D. AUTHOR'S SUMMARY

1. No attempt was made in preparing this chapter to be complete, either topically or bibliographically. Emphasis was placed on a concise review of practical information; for this reason some information is given in this summary which does not appear in the preceding portion of the chapter.

2. The only electric property of importance of iron-carbon alloys is resistivity. The value of Cleaves and Thompson (page 587), for 20°C., of 9.8 microhm-cm. for high-purity iron was selected, as was their value for the mean temperature coefficient, 0.0065 per deg. (No effort was made to report true coefficients, defined by $\alpha = 1/R_0 dR/dt$.) This unusually high value, as compared with other metals, excepting nickel and cobalt, is undoubtedly to be ascribed to the ferromagnetism of iron.

3. The electric resistivity of iron is increased by addition of carbon. Since the presence of other elements increases the resistivity also, only approximate values are of wide applicability, but such approximate values are sufficient for ordinary purposes. Other factors which affect resistivity are the form in which carbon is present and mechanical history of the specimen. A notion of order of magnitudes may be obtained from the following data, all for 20°C. and for material in the annealed condition:

Material	Microhm-cm.
Low-carbon "ingot" iron	10.7
Bessemer steel	14
Cast (carbon) steel	15
Malleable iron	32
Special cast iron	50
Ordinary cast iron	100

The resistivity of open-hearth steel such as is used in electric apparatus depends mostly on the silicon content, since carbon is held low and nearly constant.

4. Carbon steels and cast iron were once the only known magnetic materials suitable for construction of electric machinery. The displacement of carbon steel by silicon steel for trans-

former sheet did not begin until after 1900; eutectoid carbon steel was the only permanent-magnet material up to about 1910. Despite the discovery of better magnetic materials, however, carbon steels and cast irons are still used whenever their relatively low cost is sufficiently attractive; consequently the magnetic properties of iron-carbon alloys are still of practical interest.

5. The saturation magnetization of iron-carbon alloys is decreased as their carbon content is increased; the decrease is about 1400 gausses for each 1 per cent carbon, up to at least 2 per cent carbon. This decrease—somewhat more than 6 per cent—is small as compared with the change of other properties upon addition of carbon. Maximum permeability, for example, is reduced from the hundreds of thousands to less than 1000 by the addition of 1 per cent carbon. The result is that magnetic induction of iron-carbon alloys is relatively small, especially in weak fields. The coercive force of iron-carbon alloys is increased rapidly by carbon in solution and less rapidly by carbon present as free cementite. The residual induction does not change markedly with change of carbon content, at least up to 2 per cent carbon. The hysteresis loss of iron is increased rapidly by the addition of carbon. Considered as an impurity, carbon is one of the elements most injurious to the magnetic properties of iron for all purposes except the making of permanent magnets.

6. The effect of cold work on the magnetic properties of iron-carbon alloys is similar to its effect on iron, although the magnitude of the effect may be less in the higher carbon alloys. In general, cold work decreases permeability and increases coercive force. If the best magnetic properties of a given steel are required, annealing after cold work, such as rolling, is necessary. Once the sheet is annealed, the usual rule operates: don't cold work it—even shearing reduces the permeability. The best possible magnetic properties of carbon steel are seldom required however. If this material is used because of its low cost, it is probable that no manufacturer can go to the expense of annealing after shearing or punching.

7. Carbon, as was stated, becomes really useful in the making of permanent magnets. These can be and are made of heat-treated carbon steel, although the best permanent-magnet steels are of the high-alloy type.

8. Carbon is useful also in cast and malleable irons, but for properties other than magnetic. Because of their low cost and ease of casting or machining to shape, these materials are often used in spite of their inferior magnetic properties. Furthermore, the high electric resistivity of cast iron is sometimes advantageous in that eddy-current loss in alternating-current machinery is less than it would be if some of the other materials were used.

9. The mechano-, galvano-, and thermomagnetic effects in iron-carbon alloys are of little interest, consequently they may be discharged in short order. Magnetostriction is decreased by increased carbon content; *i.e.*, the maximum positive value of $\Delta l/l$ of 0.8 per cent carbon steel is about 0.5×10^{-6}. The Villari point is displaced to higher field strength by increased carbon content. There is no information on the Wiedemann effect. Reported values of the Hall coefficient range from 12×10^{-3} to 33×10^{-3}. The Righi-Leduc coefficient of steel is somewhat larger than that of iron, as is the Ettinghausen coefficient. In general, the Thomson effect is the smaller the harder is the steel. There is no useful information on the Nernst effect.

CHAPTER XVII

MISCELLANEOUS ENGINEERING PROPERTIES OF COMMERCIAL IRON-CARBON ALLOYS

Hardness—Impact Resistance—Damping Properties—Deep-drawing Properties—Machinability—Wear Resistance—Weldability—Author's Summary

Most of the physical properties discussed in Chapter XV, for example modulus of elasticity, density, thermal expansion, and thermal conductivity, can be precisely measured and expressed in fundamental units which are clearly understood and universally accepted. There are other properties of metallic materials which, although not capable of such precise measurement as the physical constants, are well understood and well standardized; tensile strength is a good example.

There remains, however, a group of properties or qualities of steel, most of which are of paramount importance in some engineering applications, which cannot readily be expressed in definite terms or in terms of some well-standardized test. Among these, discussed in the present chapter, are hardness, impact resistance, damping, deep-drawing characteristics, machinability, wear resistance, and weldability. Furthermore, numerical values for these properties, where such values can be obtained, depend to a considerable extent on the testing method chosen: it often happens that an equally logical method of testing for one of these qualities will place the materials tested in a different order. Until these testing methods become more fully standardized and their results become more definitely correlated with service tests, their evaluation is attended with such vagueness that reports of experimental work, to be intelligible, often need to be accompanied by complete descriptions of the methods used.

A discussion of these qualities usually tends to become, despite all efforts to the contrary, a discussion of testing methods, which is not the purpose of this monograph. Without attempting to deal in detail with the intricacies of the testing problem, it seems

necessary to make some mention of certain of these vague properties since the engineering choice of a steel of given carbon content or given structure, or the adoption of an alloy steel instead of a carbon steel, is often predicated on the degree to which one or more such properties may be present or lacking.

A. HARDNESS

It has been shown in previous chapters that valuable correlations exist between hardness and many other properties of steels. Tensile strength, endurance limit, machinability, wear resistance, etc., are dependent upon those inherent properties, of which hardness is a measure. The process of work hardening and that of hardening by quenching and tempering, as well as that of precipitation hardening, as these terms indicate, produce changes whose outstanding effect is to modify the hardness of the initial material.

237. Relation between Carbon Content, Brinell Hardness, and Other Properties.—That there should be a fairly direct correlation between indentation hardness, as measured by the Brinell number, and tensile strength follows from the fact that most materials have rather similar strength characteristics in tension and compression; and Brinell hardness, being measured in terms of kilograms pressure per unit of indented area of the ball impression, is, in fact, a compression value.

The correlation has been discussed by a number of observers. Abbott[31] summarized these relations for 5000 tests on 300 different steels and found straight-line relationships, with lines of slightly different slope for different classes. He concluded that a low- and a high-carbon steel, heat treated to the same strength, will have the same hardness. Greaves and Jones[179] correlated a large number of values from their own experience and from previous literature and concluded that the ratio of tensile strength (in tons per sq. in.) to Brinell hardness was 0.23 for normalized or annealed low-carbon steels, 0.22 for normalized or annealed medium-carbon steels, 0.215 for heat-treated carbon and alloy steels when the Brinell is below 250, and 0.21 for heat-treated alloy steels with Brinell values of 250 to 400. In United States units these ratios (of tensile strength, in units of 1000 lb. per sq. in., to hardness) become 0.52, 0.50, 0.485, and 0.47.

Hoyt[70] gave ratios of 0.52 to 0.58 for low-carbon, 0.48 to 0.54 for medium-carbon, and 0.50 to 0.54 for high-carbon steels. Aitchison[64] gave the ratio, in British units, as 0.22 to 0.24, *i.e.*, 0.495 to 0.53 in United States units. László[265] stated that, in metric units, the ratio varies from 0.31 to 0.41, and that for medium-hard pearlitic carbon steels it is 0.36. These values correspond to 0.445 to 0.585 and 0.51 in United States units.

The easily remembered ratio of 0.5 is a fair approximation for many purposes. Hoyt[70] pointed out that the ratio is subject to considerable fluctuation in cold-worked steels, and Greaves and Jones[179] found that the ratio is dependent on the hardness and on the yield ratio. Since all steels of the same hardness do not have the same yield strength, the ratio depends to some extent upon composition and heat treatment. (Since, as has been shown in Chapter VII, grain size affects the yield ratio, grain size should be added to the list of variables.) "For a given class of steel," Greaves and Jones stated, "the ratio decreases with increase of yield ratio and also decreases with increase of hardness up to 375 to 450, depending on the composition of the steel. In material of greater hardness, it rises again."

The approximate relation between Brinell hardness and tensile strength, as correlated from published and some unpublished data, is shown in Fig. 223 (page 619) and Table 108, page 620. In the latter, the ratio of tensile strength to Brinell varies from 0.477 to 0.485 for all of the values except the last two, in which case the ratio is almost exactly 0.5.

238. Other Indentation-hardness Tests.—In other methods of determining indentation hardness a diamond ball, cone, or pyramid is used; in some, the depth of the indentation rather than the indented area is measured. Among the latter are the Rockwell, Monotron, and Vickers methods, which are used on materials so hard that the Brinell ball is likely to be deformed, where a Brinell impression would injure the appearance of the article tested, or where the hardness of a small area is desired. The most popular and widely used of these other indentation-hardness tests is the Rockwell. This test is rapid and convenient; it measures the depth of a very small impression and, by employing steel balls of various diameters and a diamond cone, and by making use of various loads and several scales, it can be used on ferrous and non-ferrous materials of all degrees of hard-

ness. Methods of making these tests are described in the "Metals Handbook,"[821] and the American Society for Testing Materials[588, 589] has set up standards for Brinell and Rockwell testing. Cowdrey[360] has described a mutual-indentation hardness test which gives Brinell-hardness values and requires no ball. Harder and Grove[621] have applied this to high-temperature hardness testing.

Vickers diamond-pyramid values are usually reported in Brinell units, *i.e.*, as "Vickers Brinell."

239. Scleroscope and Herbert Hardness Tests.—The scleroscope method, developed by Shore and based upon the rebound of a falling diamond-pointed weight, is applicable to hard materials like case-hardened steel but is not so satisfactory for very hard ones like nitrided steel. Scleroscope hardness is sometimes considered to reflect the yield strength of the material rather than its tensile strength.

A ball hardness test in which a heavy frame, carrying a ball, is supported by the specimen to be tested and allowed to swing as a pendulum is applied in the Herbert[451] hardness tester. The time of swing is measured and is supposed to give information on the rate of work hardening—machinability and the like. The apparatus is sensitive and somewhat difficult to handle, and curves of Herbert hardness versus other properties are liable to be very jagged and to indicate the presence of considerable experimental error in the average operation of the equipment. It has not come into wide use.

240. Relation between Brinell and Other Hardness Tests and between Hardness and Tensile Strength.—Since most of the older hardness data are in Brinell units, it is important, for purposes of comparison, to know the relation between hardness values obtained in various ways. Much work has been done along this line. Wallichs and Schallbroch[349] gave Brinell-Herbert relationships as:

$$\begin{aligned}\text{Brinell} &= 24 + 0.305T_s^2\\ &= 41 - 3.9T_d + 0.78T_d^2\\ &= 239 - 17I_s + 0.52I_s^2\\ &= 27 - 4.6I_d + I_d^2\end{aligned}$$

where T is time hardness and I is induced hardness. The subscripts s and d refer to steel and diamond balls. The work-

hardening factor of the Herbert test was found to have no relationship to Brinell hardness.

A large number of investigators have studied the relation of the various hardness tests to each other. Among these may be mentioned Spalding,[140] Cowdrey,[151] Brumfield,[175] Moore,[227] Heller,[304] and Petrenko.[401] Petrenko determined the theoretical and experimental relationships between Brinell numbers and Rockwell-hardness numbers obtained with the 1/16-in. ball (100-kg. load; *B* scale), the 1/8-in. ball (100-kg. load; *E* scale), and with the diamond cone (150-kg. load; *C* scale). He plotted his results in a series of curves which indicate that conversion of the Rockwell values to Brinell may be made with an expected error not greater than 10 per cent. Conversion of Rockwell hardness to tensile strength may be made with an expected error not greater than 15 per cent.

The work of all the investigators mentioned above was reviewed by Sawin and Stachrowski,[567] who determined Brinell, Vickers-Brinell, Monotron, and Rockwell diamond hardness values on a large number of standard steels and plotted their results and those of Hessenmüller[377] in a chart for practical use at the Škoda works.

Sawin and Stachrowski summed up their results as follows:

1. The Vickers diamond hardness-testing machine is well adapted for tests of hardened steel especially if the shape of the work is complicated.
2. The Rockwell machine has its place when work of simple shape has to be tested on a production basis.
3. The Monotron machine may be of advantage when tests on a production basis have to be performed and at the same time the state of the surface has to be investigated.
4. Brinell-hardness numbers up to 500, obtained with a 10-mm. ball and 3000-kg. load, and Monotron constant-diameter values show a linear relation with Vickers-hardness numbers.
5. Rockwell *C* hardness numbers from 20 to 50 may be converted into Brinell-, Vickers-, and Monotron-hardness numbers. Outside of this range the conversion is not accurate as the relation between the Rockwell values and the other hardness values deviates from a straight line.

The results of Sawin and Stachrowski also bring out that, even with the work-hardened Hultgren ball, Brinell values above 500 are affected by deformation of the ball, which doubtless accounts for the failure of proportionality of Brinell and tensile

values at high hardness. Tungsten carbide balls are now available and in tests with these, or with the diamond indenters, the tensile-hardness relationship should probably follow the same curve as at lower hardness.

The relation between hardness values obtained by the various methods, and the relation between hardness and tensile strength

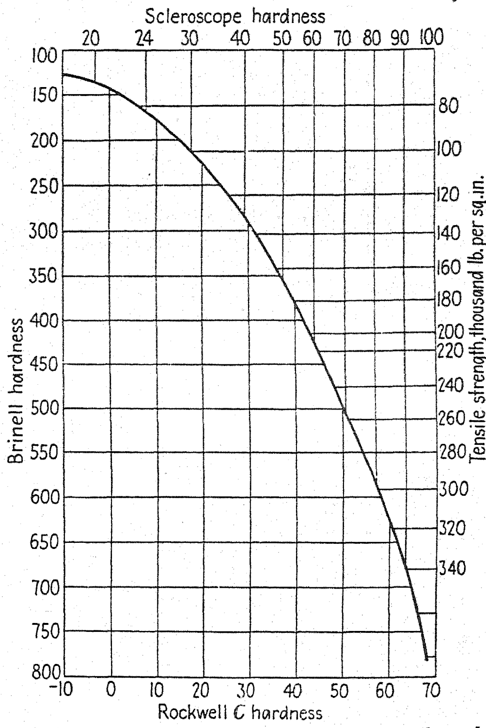

FIG. 223.—Approximate relation between tensile strength and hardness determined by various methods. (*French and Sands.*[700])

have been determined and reported by a number of investigators and tabulated in several handbooks. A typical chart of this relation, as prepared by French and Sands,[700] is shown in Fig. 223. Table 108 is a correlated summary of four published and two unpublished* tabulations. It is likely that the values

* Unpublished data, obtained by checking a large number of actual hardness and tensile tests with data from the literature, were kindly supplied for this correlation by A. P. Spooner, Bethlehem Steel Company, and J. B. Johnson of U. S. Army Air Corps, Wright Field.

plotted in Fig. 223 are not quite so accurate as the ones given in Table 108, although the former are probably sufficiently accurate for practical purposes. It is characteristic of various published hardness tables that the conversion values for Brinell to Rockwell *C* hardness as reported by various investigators check with fair accuracy. There seems to be a wider variation in the conversion to scleroscope values; the widest variation,

TABLE 108.—CONVERSION TABLE FOR HARDNESS AND TENSILE STRENGTH

Brinell, 3000 kg.		Rockwell *C*	Scleroscope	Vickers	Tensile strength, lb. per sq. in.
Diameter of impression, mm.	Number				
4.30	197	14	28	197	95,000
4.20	207	16	30	209	100,000
4.10	217	18	32	217	105,000
4.00	229	21	34	226	110,000
3.90	241	23	35	240	116,000
3.80	255	25	36	255	122,000
3.70	269	27	38	272	129,000
3.60	285	30	40	285	136,000
3.50	302	32	42	305	144,000
3.40	321	34	45	320	153,000
3.30	341	36	48	344	164,000
3.20	363	38	50	380	175,000
3.10	388	41	53	401	187,000
3.00	415	44	57	435	200,000
2.90	444	47	61	474	216,000*
2.80	477	49	65	534	238,000*
2.70	514	52	69	587	260,000*

* Uncertain.

however, is in the conversion from hardness to tensile strength. The tensile-strength values in the four published and the two unpublished tables mentioned on the previous page varied so widely for hardness below 197 and above 514 Brinell that no correlation could be made for Table 108.

The relation shown in Fig. 223 and Table 108 is valid only for carbon and low-alloy steels; it is not valid for high-speed and other highly alloyed steels, for case-hardened materials, or for non-ferrous alloys.

The literature on hardness testing is extensive. O'Neill dealt with the subject in detail in a number of papers(192,275,399) and in a recent book;(745) Hankins(301) summarized the literature of the last decade; and Mellor's(741) discussion includes many recent references.

This brief survey of indentation-hardness testing should not omit mention to Meyer's work. In 1908, Meyer(13) reported the results of an investigation of the principles underlying the ball-indentation hardness (Brinell) test. He showed that, provided the loading was maintained long enough for equilibrium to be reached, the following relation is valid:

$$P = ad^n$$

where P is the load in kg., d is the diameter of the impression in mm., and a and n are constants for the material under test. Meyer showed that the resistance which a material offers to deformation varies with the degree of deformation, and cannot, therefore, be expressed by a single number. A single hardness value can be given only when it is qualified by a statement of the specific amount of deformation used in securing that value; it represents, therefore, only a single point on the hardness-strain curve.

As the whole stress-strain curve in tensile testing is more valuable than a single point on the curve, so the whole hardness-strain curve is more valuable than a single Brinell number. By means of the Meyer formula the whole curve is known if the constants a and n are known. The distinction between the value of a single point and the whole curve is apparent if it becomes necessary to compare the indentation hardness of two different metals. Meyer showed that the hardness of soft iron, for example, is less than the hardness of copper for small indentations; at an indentation value of $d = 0.436$ the two metals resist indentation equally well (are of the same hardness), but for larger indentations the iron is harder than the copper. This is due to differences in the hardenability or n values for the two metals. An excellent summary and discussion of Meyer's work were given by Hoyt(123) and more recently in O'Neill's book.(745) O'Neill gave references to many of Meyer's investigations on the principles of hardness testing, reported since his original paper in 1908.

241. Scratch-hardness Tests.—For cases where the hardness of microscopic areas is required, some modification of the old Turner-Martens sclerometer is used. In this form of hardness determination a sharp point, as of a diamond, is drawn across the specimen under a definite load and the width of the scratch measured microscopically. The instrument mostly used for this purpose in the United States is the so-called "microcharacter," described by Bierbaum,[358] who stated that there is no direct relationship between scratch hardness and indentation hardness since the indentation methods bring into consideration more than one microconstituent. He prefers to use the reciprocal of the square of the width of the scratch as the hardness number.

Briggs and Williams[702] found disadvantages in this method of recording the numbers and proposed to use the width in microns as the hardness number. In their practice, the microcharacter is used with either a 3-g. or a 6-g. weight on the diamond point which is lubricated, preferably with a light machine oil. By calibration on single-phase materials of a range of hardness they found that straight-line calibration curves result from a comparison of scratch width with Rockwell E and B scales, when a 3-g. weight is used, and with the Rockwell C scale, when a 6-g. weight is used. Such curves are:

$$\begin{aligned} &\text{zero } R_E = 45M_3; &\quad 100R_E &= 5M_3 \\ &\text{zero } R_B = 17M_3; &\quad 100R_B &= 3.6M_3 \\ &\text{zero } R_C = 12.6M_6; &\quad 65R_C &= 2M_6 \end{aligned}$$

Thus they were enabled to report their scratch-hardness values in terms of Rockwell hardness. The equivalent Rockwell hardness of the ferrite constituent of pearlitic carbon steels, annealed and furnace cooled, was found to be as follows:

Carbon, Per Cent	Equivalent Rockwell C Hardness
0.008	11.1
0.20	14.6
0.40	17.8
0.60	20.2

The American Society for Testing Materials[147] reported the results of a cooperative investigation in which seven observers made scratch-hardness tests on the same series of specimens.

The results were erratic. The limitations of scratch-hardness testing are shown further by work recently reported by Scheil and Tonn.[765] Cold working soft irons increased the Brinell hardness from 100 to 200 while the scratch-hardness values were hardly affected. In steels of varying carbon content, in the spheroidized condition, curves for Brinell and scratch hardness were parallel. When these steels were quenched from 1100°C. (2010°F.) and not tempered, the Brinell hardness increased with carbon content until the latter reached 1 per cent, above which, due to retained austenite, the hardness was lower. Scratch-hardness values increased with carbon content until the latter reached 0.7 per cent; beyond that percentage the scratch-hardness values remained the same. This method, thus, does not reveal the presence of retained austenite. Neither does it, according to Scheil and Tonn, reveal the same increase in hardness in precipitation-hardened alloys as is shown by the Brinell test. From the scatter of points in Scheil and Tonn's scratch-hardness curves it is apparent that this method is not reliable in determining the hardness of quenched high-carbon steels.

B. IMPACT RESISTANCE

Low-carbon steel may be so tough that it can be battered with a sledge without breaking; quenched high-carbon steel may be so brittle that it shatters when dropped. The elongation and reduction of area in the static tensile test give a general idea of toughness versus brittleness, and the area under the static stress-strain curve gives an even better one. But under some conditions a steel which would be classed as tough by static tests may, by other tests, be classed as brittle.

It is well known that the way to break a tough metal is to notch it, thus concentrating the application of stress, and strike it a sharp blow. In service, stress concentration may occur at sharp reentrant angles or poor fillets resulting from improper design or from the exigencies of design; tool marks and other notches may also be present. In many types of service sudden shocks may be intentionally or accidentally applied.

242. Relation between Notch Brittleness and Strength.—The effort to evaluate the behavior of steel specimens in which there is a concentration of stress has led to the study of notched

bars in the static tensile test. Kuntze[264,540,541] made tests on a series of tensile bars, each with the same diameter but turned down with a V-groove to different depths so as to leave an increasingly smaller cross-section at the base of the notch. The flow of the metal above the yield strength is more and more restricted as the notch becomes deeper, and the tensile strength calculated on the minimum section is greater. By plotting the data for different minimum sections, extrapolation may be made to a theoretical zero cross-section[751] (see Fig. 224). The tensile strength found by this extrapolation is termed "Trennfestigkeit" or technical cohesion strength. It is thought

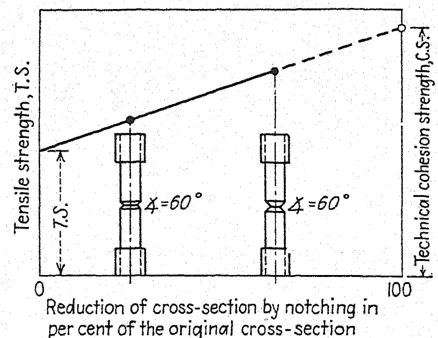

FIG. 224.—Representation of Kuntze's method of determining technical cohesion strength (Trennfestigkeit). (*Piwowarsky.*[751])

by Kuntze that this is a fundamental property of metals and that the ratio between it and the ordinary tensile strength on unnotched bars gives information on notch sensitivity. The method is in use at the Staatliches Materialprüfungsamt, Berlin-Dahlem, but has so far found relatively little use elsewhere; insufficient data are available to permit evaluating the behavior of carbon steels as a class by such a test. Very rapid loading is not used with this method.

In the tensile-impact test, a notched or unnotched bar resembling the standard tensile specimen is used. The bar is broken by shock loading. Langenberg[59] has studied this test over a long period at Watertown Arsenal; but again, insufficient data are available for evaluating by its means the properties of carbon steels as a class. Mann,[807] also at Watertown Arsenal, has carried Langenberg's work farther and has investigated the effect

of variable speed upon the tensile-impact value. Mann believes that "the impact [tensile impact] at variable speeds, in combination with the static-potential curve, should completely determine the dynamic and static properties from which subsequent behavior of a material under known service conditions can be definitely predicted."

243. Single-blow Notched-bar Impact Testing.—Most impact testing is done by bending a notched bar or by breaking such a bar with a single blow applied transversely. Various types of machines and specimens have been used, but little or no standardization has been effected. In a symposium on impact testing held by the American Society for Testing Materials,(173) it was concluded that values obtained by different types of machines are not comparable. Thus, an accurate comparison of the toughness of different materials as evaluated by different types of specimens is impossible. Although impact values are not accurate enough to detect slight differences in toughness, there is little difficulty in separating tough and brittle materials by any of the tests.

One of the weaknesses of the impact test has been that the same material would at times, and apparently under the same testing conditions, give duplicate values for impact resistance which showed wide variation. This has led to considerable reflection upon the accuracy and value of single-blow notched-bar impact as a test for toughness. It has been found, however, that these erratic results are due to the material and not to the method of testing.

Some materials may have two ranges of resistance to shock at a notch; in one, the range of high resistance, the fracture is fibrous, tensile deformation and plastic flow having occurred; in the other, the low range, the fracture is granular, and failure occurs by shear with little deformation. If a material is well above the dividing line between the ranges of high and low impact, values obtained on standard bars should show high impact resistance, and duplicate tests should check closely; if a material is of such character that it is well below the dividing line, the fracture is granular and the impact resistance is low. Duplicate tests should check closely.

If, however, as occasionally happens, a material is so constituted that it is on the dividing line between the high and the

low range, a slight increase in the speed of the blow, the sharpness of the notch, a change in the size of the section back of the notch, a slight variation in temperature, or in composition, or in heat treatment may radically affect the resulting impact value, or may even cause results to vary widely on apparently duplicate tests of the same material.

Fettweis[294] made a very thorough study of the variables in the impact test and collected a bibliography of 700 references

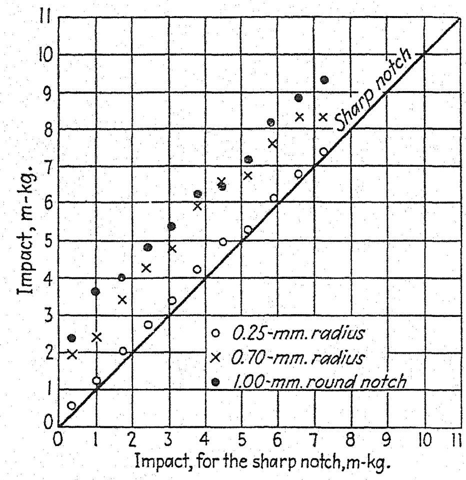

FIG. 225.—Effect of rounding the base of the notch on impact resistance. (*Fettweis.*[294])

on the subject. His Fig. 225 shows the effect of a round versus a sharp notch.

Dejean and Gerszonowicz[685] recently summarized the subject of impact testing; Fig. 226 from their paper shows the effect of specimen size for a particular steel. In commenting on this section, A. P. Spooner* wrote that reference should be made to work now being done on correlating impact values, obtained on specimens of different size, with resistance to impact in actual service:

It is felt by some that by varying the size of the specimen a measure of the ability of the steel to propagate sudden impact loads can be relatively determined. For example, if a specimen one-half the usual size has an impact value of 8 ft-lb., a normal-sized specimen 16 ft-lb.,

* Private communication.

and a double-sized specimen 32 ft-lb., the steel would be considered satisfactory. If, however, these values were, respectively, 12, 16, and 17 ft-lb., it would be considered a poor material to withstand suddenly applied loads, especially if the section in service is not uniform.

Spooner added that not enough work has been done as yet to prove or disprove conclusively this correlation.

The size of the specimen is in practice fixed by the capacity of the impact machines available. Most of these have a capacity

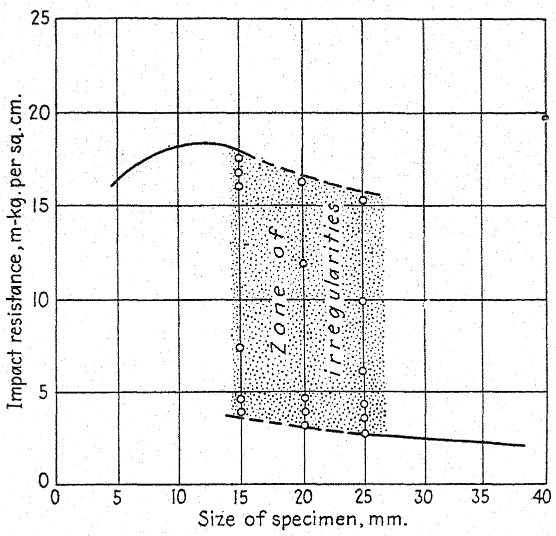

FIG. 226.—Effect of width of specimens on impact resistance. (*Dejean and Gerszonowicz.*[685])

of 120 ft-lb., on which materials of the maximum toughness usually encountered can be broken if the specimens are notched bars 10 × 10 mm. sq., or of equivalent circular section before notching. Larger machines are in use but are relatively rare.

244. Izod and Charpy Impact.—The two most common methods of impact testing, in the Charpy or the Izod machine, are so well known that only a brief résumé of their respective advantages and disadvantages is justified in this monograph.

The Izod specimen, widely used in the United States[173,821] and England,[55] is 10 × 10 mm. (0.394 × 0.394 in.) sq. with a 45-deg. V-notch 2 mm. deep and having a radius of 0.25 mm. at the bottom of the notch. The remaining cross-section is 10 × 8 mm. Round Izod specimens with a notch of the same type are

also used. The specimen is gripped at one end and the free end struck by a swinging hammer.

The Charpy specimen is supported at each end over a span in such a way that the notch is at the center of the span. On account of this method of support, it does not have to be gripped and may be very rapidly laid in place so that specimens heated or cooled for determination of impact resistance at other than room temperature may be tested easily without appreciable change in temperature. Thus the Charpy test avoids many difficulties of the Izod test when employed at other than room temperature.

The radius at the base of the notch is an important factor. The maintenance of notching tools, so that they give the standard notch, and the careful machining and checking of the finished specimens, are troublesome and costly. When a square bar is used for the Charpy test, the notch can be made comparatively easily by drilling and then making a saw cut into the hole, giving the "Mesnager" or "keyhole" notch.

This has led to rather general preference for the square Charpy bar with the keyhole notch, though the V-notch is more severe and may therefore be chosen in particular instances. It sometimes happens that a standard square Charpy bar cannot be machined from the material it is desired to study. A case in point is the determination of impact properties at room temperature on a 0.505-in.-diameter tensile bar after a creep test at elevated temperature, to show whether any embrittling action has occurred. In this case, the only standard bar which can be made from the material available is the round Izod specimen.

The size of the drilled hole for the keyhole notch in the square Charpy bar varies in different countries; in the United States[821] and in France[685] it generally has a diameter of 2 mm. (No. 47 drill). It is located so that the remaining cross-section of the bar is 10 $\times$ 5 mm. The German Charpy specimen, with drilled or cut notch, usually has an effective cross-section of 10 $\times$ 7 mm.

When the notch must be ground instead of drilled, as in very hard materials, the shallow notch has an advantage. An attempt to reconcile results of tests obtained with these different specimens is often made by reporting values in m-kg. per sq. cm. or ft-lb. per sq. in. rather than in m-kg. or ft-lb. But since the

values from different bars are not interchangeable in this way, it is preferable to describe the bar and the notch and give the values in ft-lb. or m-kg.

245. Value of the Single-blow Notched-bar Impact Test.—As has been stated before and will be emphasized again here, impact values obtained by the different machines are not comparable, nor can slight differences in toughness be detected by the usual impact test. Nevertheless, regardless of the difference in specimen or of the type of notch, and irrespective of the difference in units in which the impact value is reported, one is seldom in doubt whether the value obtained by the test represents a tough or a brittle steel. Because of the limitations of the test, results are never used directly in design. As Greaves[372] put it, the notched-bar test is an "exclusion test." An impact value for a given class of steel for a particular service may be fixed on the basis of satisfactory experience. The exclusion of material not meeting that standard is a safeguard against failure that might be met were more notch-brittle material allowed. In investigative work, the results are used for comparative purposes only, and in most cases the same comparison will result whether one or another type of bar is used.

The use of the single-blow impact test has increased greatly in recent years. Makers of ordnance, automobiles, and aircraft have found that materials of good impact resistance in laboratory tests have given good service in applications involving shock loading. Carefully made impact specimens should give results in satisfactory agreement if the material is not such as to be in a transition zone (see page 625). Lightner and Herty[542] have even drawn conclusions from variations in impact values which most testing engineers would not so long ago have considered to be within the experimental error.

Cold working reduces the impact value. The impact test is used, on specimens which have been moderately cold worked and aged, as a sensitive test for susceptibility to aging. It is used as a criterion of grain size and as a method of evaluation of temper brittleness, indicated by low impact values on slow cooling of some alloy steels after tempering. As this phenomenon is rarely encountered in plain carbon steels—although it is common in some alloy steels, especially the nickel-chromium steels—it is not discussed here.

Davenport, Roff, and Bain[684] used very effectively single-blow impact on unnotched round bars to detect microscopic cracks in hardened steel and to trace the relation between these cracks and the austenitic grain size.

246. Notched-bar Impact Tests of Carbon Steels.—In connection with the correlation of data on tensile properties in previous chapters, the effect of heat treatment, cold working, high and low temperatures, grain size, and other variables on the impact resistance has also been dealt with. There are, however, a few more data from tests in which impact resistance was the only value determined.

Körber and Pomp[162] made detailed studies of five carbon steels; four of them (*A* to *D*) were tested as rolled, as overheated for 2 hr.—to produce a large grain size [1250°C. (2280°F.) for the 0.05 per cent carbon steel and 1000°C. (1830°F.) for the others]—and furnace cooled, and after water quenching from above A_3 followed by tempering at 650°C. (1200°F.) and air cooling. A high-carbon steel (*E*) was tested as "annealed" and after oil quenching from 780°C.(1435°F.) followed by tempering at 650°C. (1200°F.) and air cooling.

The impact tests were made on 15 × 15-mm. sq. Charpy specimens, using a 120-mm. span and a 3-mm.-diameter keyhole notch. The effective section back of the notch was 15 × 8 mm. The steels used by Körber and Pomp had the following composition:

Steel	Composition, per cent				
	C	Mn	Si	S	P
A	0.05	0.33	0.02	0.028	0.020
B	0.23	0.73	0.02	0.040	0.028
C	0.40	0.78	0.17	0.043	0.019
D	0.58	0.72	0.18	0.050	0.025
E	1.00	0.34	0.25	0.021	0.011

The room-temperature tensile and impact properties are shown in Fig. 227 and the elevated-temperature impact values, after the various treatments, in Figs. 228 to 232.

It will be noted that the general trend for the impact resistance is to go down as the strength goes up. The impact is roughly

related to static ductility, but a variable such as grain size vastly affects the relation which would otherwise be expected. This

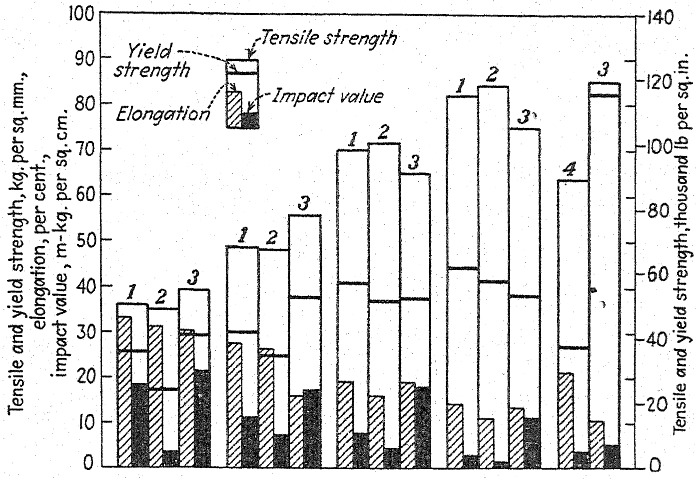

FIG. 227.—Tensile and impact properties, determined at room temperature, of five carbon steels (see page 630 for composition and treatment) as (1) rolled; (2) overheated; (3) water quenched and tempered; (4) annealed. (*Körber and Pomp.*[162])

has also been clearly brought out by the data of Herty and McBride discussed in Chapter XII.

Epstein[520] determined the impact resistance at temperatures between −80 and +800°C. (−110 and +1470°F.) of ingot iron and four low-carbon structural steels of the following analyses:

Steel	Composition, per cent			
	C	Mn	S	P
Bessemer	0.08	0.47	0.035	0.102
Open-hearth	0.21	0.37	0.036	0.008
Duplex	0.22	0.54	0.030	0.017
Izett	0.13	0.58		

Results are plotted in Fig. 233.

Impact resistance is frequently used in an attempt to detect differences between steels made by various processes, and between ordinary low-carbon structural steels and such non-

aging materials as the Izett steel. Epstein's results on these steels are also discussed in Chapter XII.

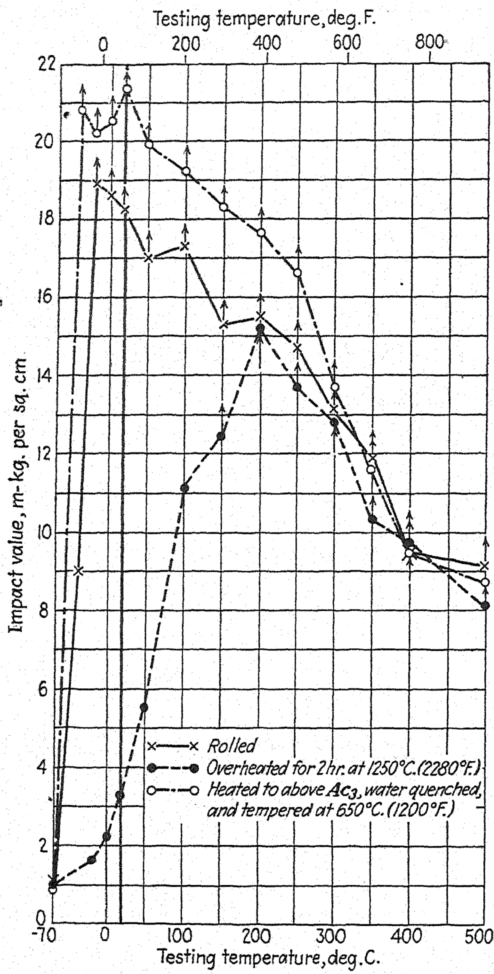

FIG. 228.—Effect of temperature on the impact resistance of steel *A* containing 0.05 per cent carbon. The arrows denote that the specimens bent but did not fracture completely; hence, true values may be somewhat higher than shown by the plotting points. (*Körber and Pomp.*[162])

247. Impact Tests of Relatively Brittle Materials.—The most difficult problem in impact testing is the evaluation of slight differences in toughness in brittle materials. The transverse

test shows such differences, when they are present in cast iron, by the values for deflection; the values obtained in a notched-bar impact test made in the usual manner are too small (a fraction of 1 ft-lb.) to reveal these variations. Likewise there are marked differences in the brittleness of high-carbon tool steels, as evi-

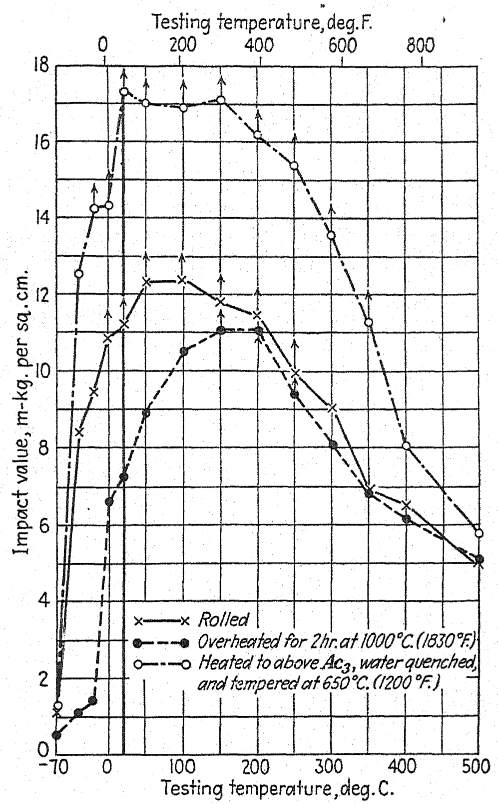

FIG. 229.—Same as Fig. 228 for steel *B* containing 0.23 per cent carbon. (*Körber and Pomp.*[162])

denced from their behavior in service, which cannot be measured by single-blow notched-bar tests.

The question of impact tests on cast iron has been considered in detail by a Committee of the American Society for Testing Materials,[586] which concluded that a drop test on an unnotched 1.20-in.-diameter bar, supported over a span of about 6 in., consisting of dropping a hammer of about 25 lb. weight upon the

bar from heights which are increased by some constant increment, for example 1 in., until fracture occurs will probably be the best evaluation of toughness. The conclusions of the Committee on these drop tests and on the accuracy of the single-blow

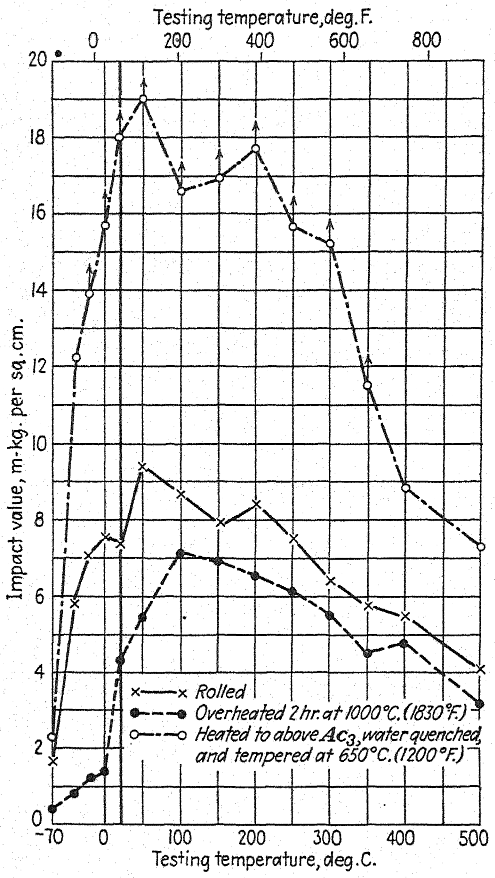

FIG. 230.—Same as Fig. 228 for steel *C* containing 0.40 per cent carbon. (*Körber and Pomp.*[162])

notched-bar impact test for determining the toughness of cast iron are given on page 298.

Luerssen and Greene[632] have developed a torsion-impact test for tool steels. Results of such tests on unnotched bars, compared with those obtained in standard Izod tests, on a steel of 1.06 per cent carbon, 0.20 per cent manganese, 0.16 per cent

silicon, 0.01 per cent phosphorus, 0.012 per cent sulphur, 0.03 per cent chromium, and 0.04 per cent nickel are shown in Fig. 234.

The test is so new that it is not yet certain whether the peak in toughness at a tempering temperature of 175°C. (350°F.), as indicated by the torsion-impact results, completely represents

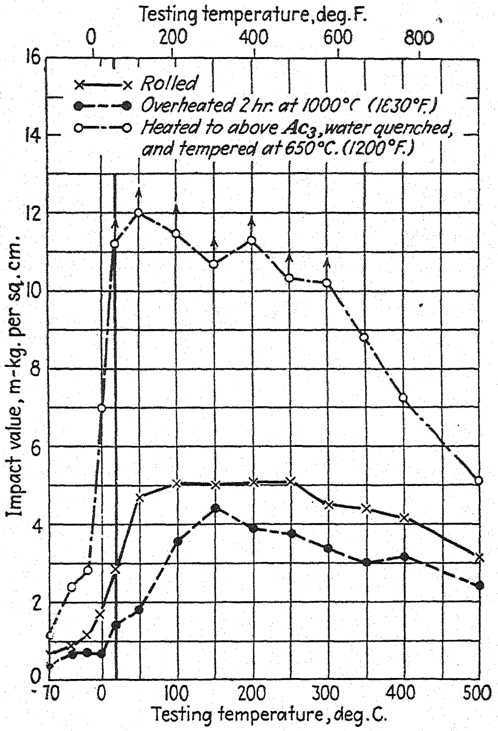

FIG. 231.—Same as Fig. 228 for steel *D* containing 0.58 per cent carbon. (*Körber and Pomp.*[162])

the differences found in service or whether this, like the results of some other impact tests, is affected by variations in size and shape of the specimen and other variables.

Scott[770] did not find this peak by the ordinary impact test, and Greene and Luerssen pointed out that it is detected only in unnotched specimens. Itihara[535] has also devised a torsion-impact apparatus.

248. Repeated Impact Tests.—Various forms of a repeated-impact test on notched bars and machines for making them, such

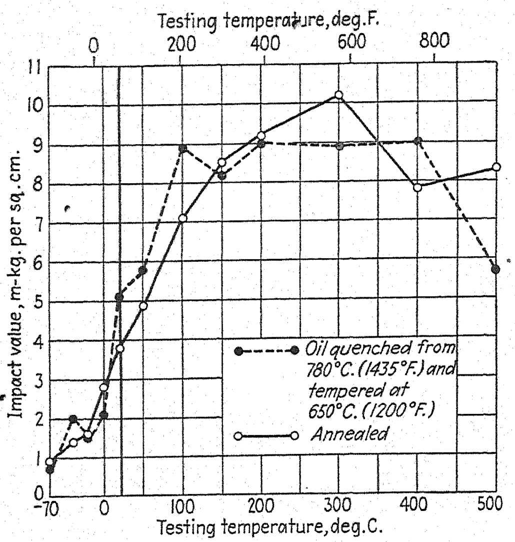

FIG. 232—Same as Fig. 228 for steel E containing 1.00 per cent carbon. (*Körber and Pomp.*[162])

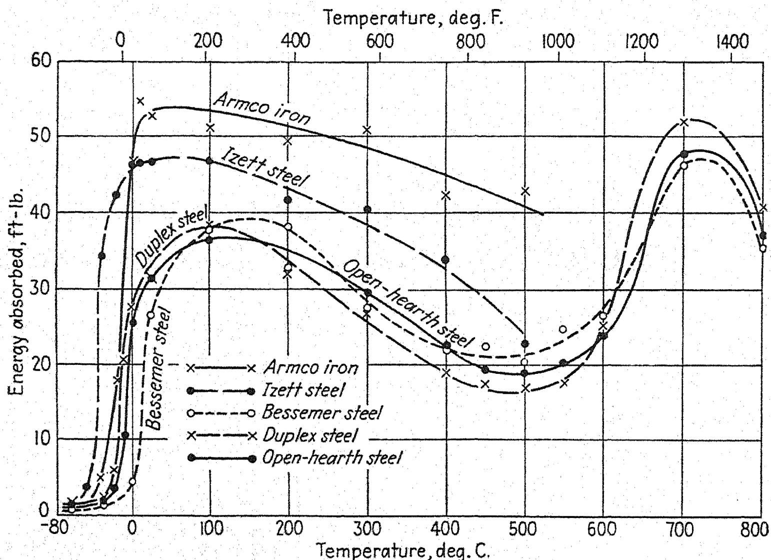

FIG. 233.—Change in notched-bar impact resistance with temperature of ingot iron and low-carbon structural steel. (*Epstein.*[520])

as the Stanton and Krupp machines, have been suggested. On the whole these tend to give much the same type of information as the single-blow impact test when the intensity of the blow is great enough to break the specimen in a few thousand blows. If it is low enough so that fracture occurs only after hundreds of thousands of blows, the information is of the same general nature as that obtained in endurance tests. Intermediate intensities,

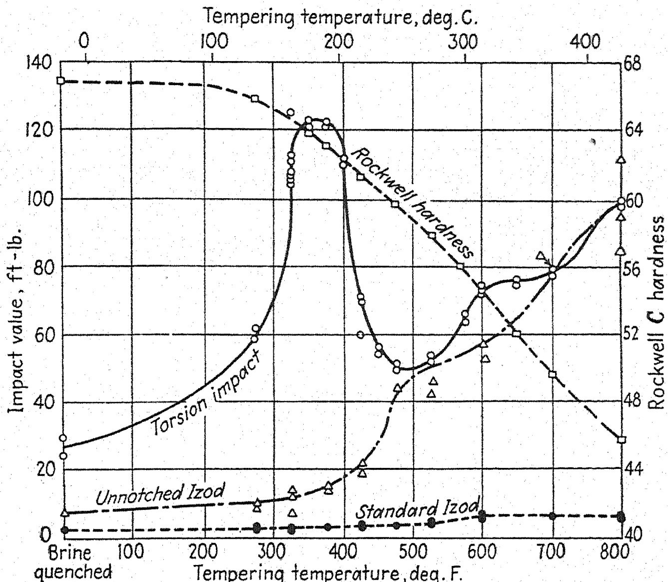

FIG. 234.—Relation between Izod-impact and torsion-impact tests on 1.06 per cent carbon steel quenched in brine from 790°C. (1450°F.) and tempered as shown. (*Luerssen and Greene.*[633])

producing fracture after 10,000 to 20,000 blows, give results difficult to interpret. The test has been found useful by some metallurgists but, on the whole, seems to be falling into disrepute and no useful purpose would be served by assembling here such data as are available on carbon steels.

C. DAMPING PROPERTIES

Damping is the absorption of energy within a material. Were there no dissipation of this energy, a tuning fork set to vibrating in a vacuum would continue to vibrate. The stress-strain dia-

gram within the elastic limit is not a single straight line but actually an hysteresis loop, whose width is a measure of the damping or dissipation constant of the material. Ordinarily this deviation from a straight line is so minute that stress-strain curves cannot be used practically to determine damping. Walther[781] pointed out that the amplitude of motion involved may be extremely small so that the actual energy loss is almost infinitesimal. An example cited by him relates to vibration under stated conditions of a lead bar, which reveals that damping is occurring, but the energy involved is so small that, if the bar were perfectly insulated, the vibration would have to be maintained for 3 million years to raise the temperature of the bar 1°C. Yet the energy absorbed by lead is very high compared with other metals.

249. Damping Capacity.—Damping capacity, according to von Heydekampf,[456] was defined in 1923 by O. Föppl as "the amount of work dissipated into heat by a unit volume of the material during a completely reversed cycle of unit stress." It is measured in in-lb. per cu. in. (or cm-kg. per cu. cm.) per cycle. For comparison of the damping capacity of various materials the *specific damping capacity* is used. This is the damping capacity divided by the potential energy at the maximum stress of the cycle. The resulting value is a dimensionless quantity.

Within the last 25 years damping, especially in connection with machine tools and internal-combustion engines, has received considerable attention. In one of his papers von Heydekampf[456] cited 38 references, most of them to reports by the Wöhler Institute, where much of the research on this property has been done.

The importance of damping capacity to engineers is obvious. For many industrial applications—for example a bed for a high-speed machine or a crankshaft for an internal-combustion engine—a material with high tensile strength but low damping capacity, which permits vibrations to build up to a high magnitude, may fail much sooner than a material with low tensile strength but high capacity to "damp out" vibration.

Damping capacity may be measured by a number of methods. These have been reviewed briefly by von Heydekampf.[456] The most common is the Föppl-Pertz machine, developed at the

Wöhler Institute,* which measures free vibration damped only by the internal friction of the material. Typical curves, taken with this machine, of the damping capacity of cast iron, ball-bearing steel, non-ferrous alloys, and rubber are shown in Fig. 89, page 282. Pohl(651) has also published damping curves for cast iron, carbon steel, and duralumin.

By plotting the specific damping capacity, in per cent, against maximum torsional fiber stress, in lb. per sq. in., a series of curves is obtained by which the relative ability of a material to damp out vibration can be determined. Von Heydekampf(456) gave such curves for a number of ferrous, non-ferrous, and other materials. The composition of some of the ferrous materials used is as follows:

Mark	Composition, per cent							
	C	Mn	Si	Ni	Cr	V	Mo	W
Cast iron	Pearlitic cast iron							
Low carbon	0.22	0.04	0.07					
High carbon	0.93	0.10	0.11					
Ni-Cr-Mo	0.16	0.40	0.15	4.90	1.90	....	0.70	
Ni-Cr-1	0.3	....	....	4.2	0.95			
Ni-Cr-2	0.3	....	....	4.2	0.95			
Ni-Cr-W-Mo	0.4	0.45	0.3	2.5	1.15	....	0.25	0.9
Cr-V	0.5	0.6	0.2	....	1.05	0.17		

The specific damping capacity for these materials is plotted in Fig. 235. It will be noted that the highest damping capacity is exhibited by the cast iron, and that low-carbon steel ranks next. Most of the alloy steels tested had a very low capacity to damp out vibration. The high damping capacity of cast iron and low-carbon steel (case-hardened) as compared with some alloy steels and with red brass is also shown by Fig. 129, page 406.

250. Acoustic Properties of Carbon Steels.—The acoustic properties of a metal are closely related to the damping capacity; the duration of sound from a vibrating body is that portion of a damping graph in which the vibrations are audible. The

* The Föppl-Pertz machine has been described by a number of the authors given in von Heydekampf's(456) bibliography. A brief description also appears in *Metals & Alloys*, v. 2, No. 2, 1931, p. 28.

acoustic properties of cast iron and carbon steels were studied by Robin 25 years ago, but relatively little attention was paid to his work and to its relation to damping until the last 10 or 12 years.

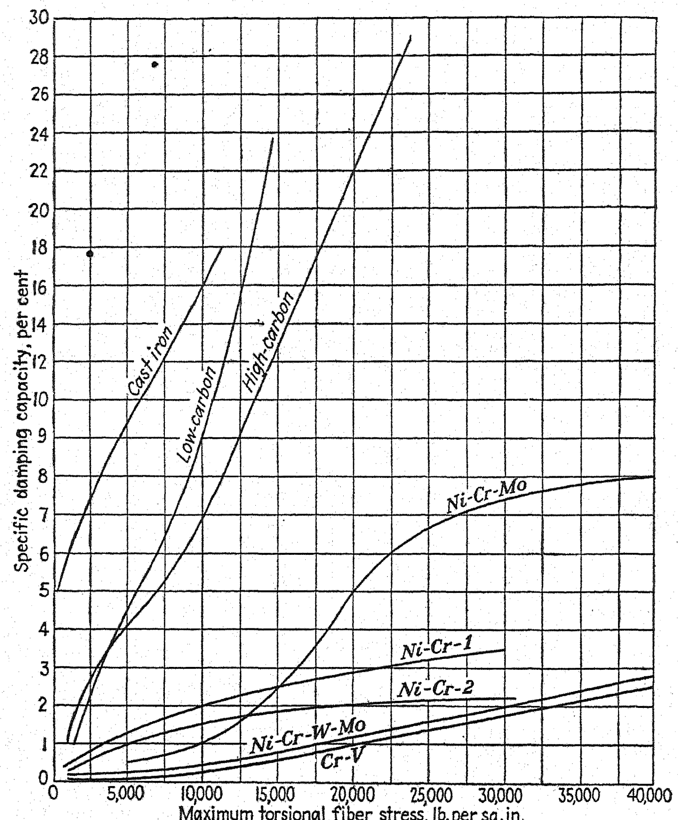

FIG. 235.—Relative damping characteristics of some commercial ferrous alloys. Note change of scale at 10 per cent damping capacity. See page 639 for composition of materials. (*After von Heydekampf.*[456])

Robin[21] investigated the effect of elevated temperatures on the duration of sound caused by striking iron and steel bars 15 mm. in diameter and 200 mm. long. Some of his results are shown in Figs. 236 and 237. The behavior of hardened steel is exemplified by Fig. 238 of a quenched and tempered eutectoid steel. The annealed specimens were "dead" at about 120°C. (250°F.) but became resonant again at higher temperatures,

while the hardened steel showed a curve of the same general slope only after tempering at 550°C. (1020°F.). The phenomena reported by Robin are difficult to correlate with other properties.

Pomp and Zapp[652] studied the acoustic properties of a series of quenched and tempered carbon-steel rods. Some of their results are shown in Figs. 239 and 240. They also determined the damping properties on a few specimens and correlated damping

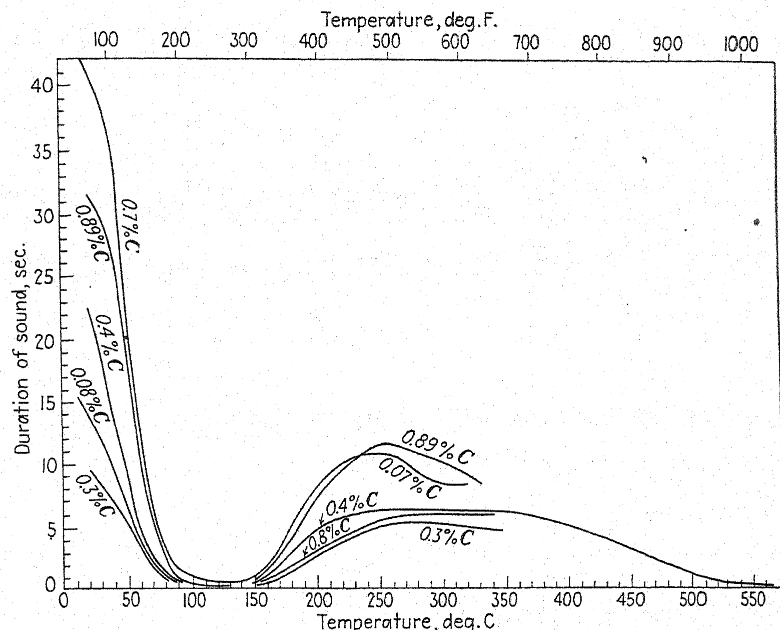

FIG. 236.—Duration of sound in annealed steels containing 0.07 to 0.89 per cent carbon. Test specimens were 15 mm. (0.59 in.) in diameter and 200 mm. (7.87 in.) long and were suspended for ¼ of their length. (*Robin*.[21])

with acoustic properties. It was found that the steels with the highest damping capacity made the poorest gongs. The sonorousness and the damping were found to be related to the metallographic structure and the internal stresses present.

251. Relation of Damping to Structure.—Hempel and Plock[800] made tensile, endurance, and damping tests on a series of carbon steels. The damping was determined by varying the amplitude and plotting this, expressed in percentage, against the logarithmic decrement. This latter value is a measure of the rate at which vibration dies down; an increasing logarithmic

decrement indicates an increasing damping capacity. No definite relation between carbon content and damping capacity was found.

To determine the effect of structure, two steels containing 0.39 and 0.77 per cent carbon were used. These were treated so that various structures were obtained, including completely troostitic, sorbitic, and lamellar pearlitic structures. In general it was found that the specimens having structures corresponding

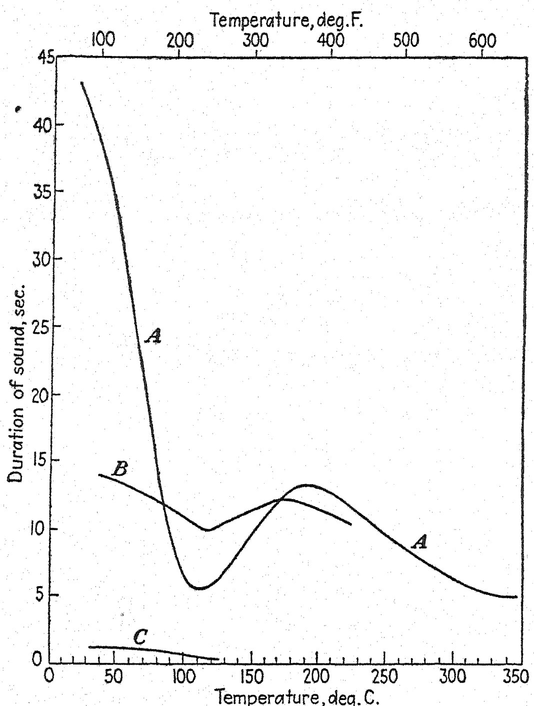

FIG. 237.—Same as Fig. 236 for A—1.2 per cent carbon steel; B—white cast iron; and C—gray cast iron. (*Robin.*[21])

to high static strength also had high endurance (alternate tension and compression) and low damping capacity. Hempel and Plock's values for the 0.39 per cent carbon steel are reproduced in Fig. 241. Troostite has the lowest damping capacity; furthermore, the damping capacity of a steel which is wholly troostitic changes but slightly as the amplitude increases. The superiority

of the pearlitic structure in damping out vibration becomes more clearly apparent as the amplitude increases.

This investigation is further confirmation that, in general, coarse-grained low-strength steels are more suitable for applications in which rapid damping of vibration is important.

252. Relation of Damping to Other Properties.—Von Heydekampf(456) has shown that the damping capacity is reflected with

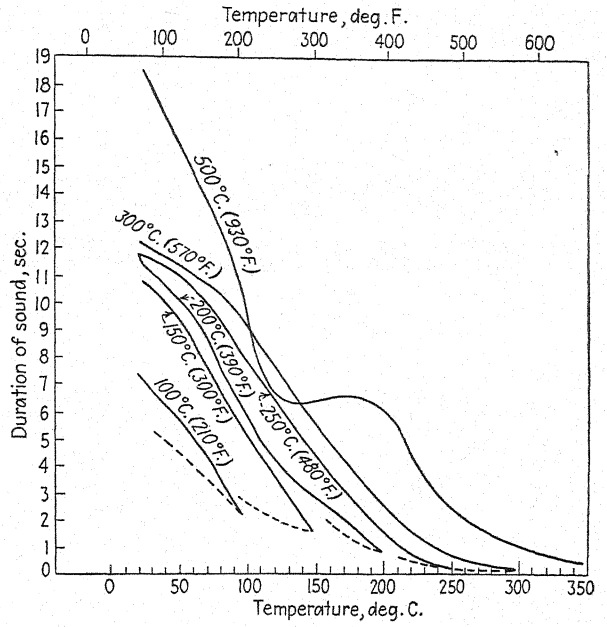

FIG. 238.—Same as Fig. 236 for 0.89 per cent carbon steel, quenched from 900°C. (1650°F.) and tempered as indicated on curves. (*Robin.*(21))

fair accuracy in the hysteresis curve of the reversed-bend test. Thus it will vary as the plastic deformation varies. This in turn varies inversely with the relative modulus of elasticity. In other words, the stiffer the material the lower its damping capacity.

In none of the work at the Wöhler Institute, reviewed by von Heydekampf, has any direct relation been detected between damping capacity and fatigue. Damping curves show no "knee" or other irregularity which indicates a fatigue limit. Thus, damping tests cannot be used as reliable short-time fatigue tests. (This is discussed briefly on page 406.) On the other hand, a

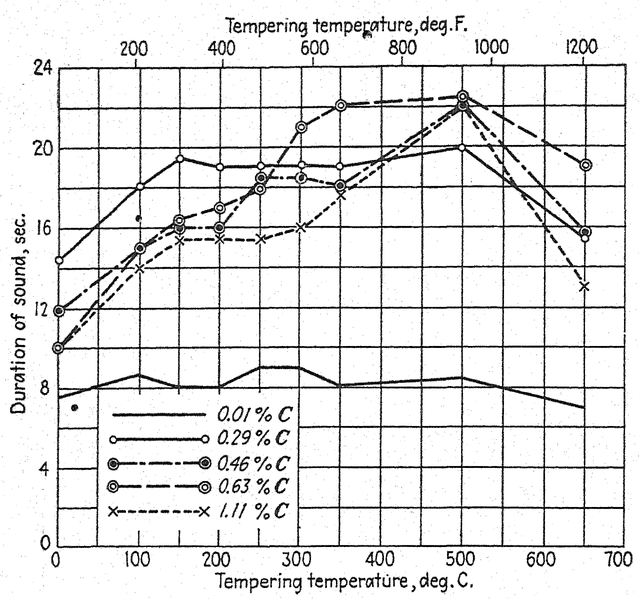

FIG. 239.—Effect of tempering for 15 min. on the duration of sound in quenched carbon steels. (*Pomp and Zapp.*[652])

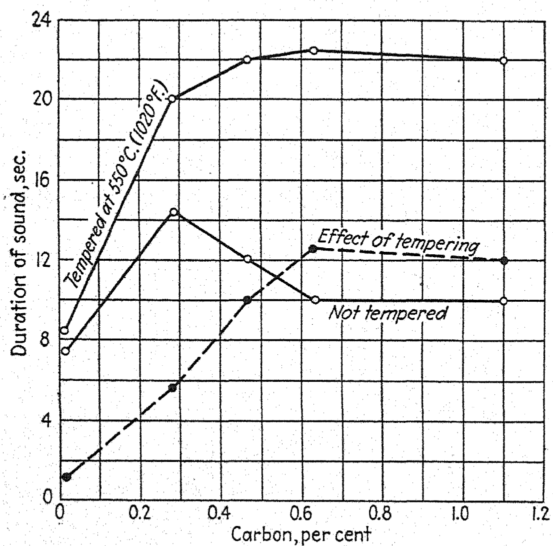

FIG. 240.—Effect of carbon content on the duration of sound in quenched and quenched and tempered carbon steels. (*Pomp and Zapp.*[652])

damping curve is a sort of stress-strain diagram but is much more accurate than the ordinary static stress-strain plot and measures extremely small amounts of non-elastic deformation, much smaller than can be measured by the ordinary extensometer. From the curves shown in Fig. 235 it is evident that there is no true proportional limit. Damping tests, therefore, again confirm the fact that Hooke's law is only an approximation.

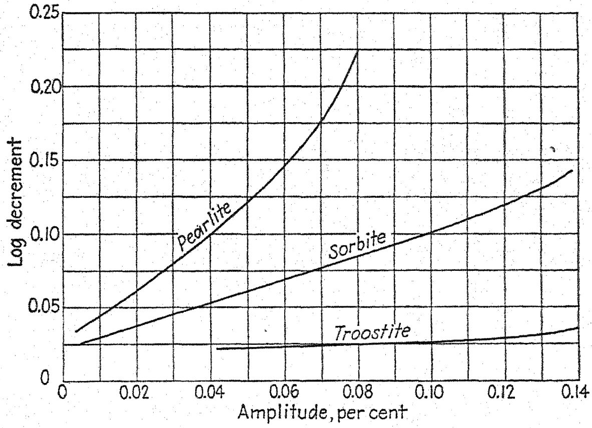

FIG. 241.—Effect of structure on the damping capacity of a 0.4 per cent carbon steel. (*Hempel and Plock.*[800])

Von Heydekampf believes that damping values can be coordinated with fatigue strength in the same way that elongation is coordinated with tensile strength. As elongation is a measure of static ductility, so damping may be made a measure of dynamic ductility. Thus, chromium-vanadium steel, having low damping capacity (Fig. 235), is dynamically brittle, while cast iron, a statically brittle material, is dynamically ductile.

There may be a fairly close relationship evolved between notched-bar impact resistance and damping capacity. According to von Heydekampf,

damping capacity seems to be related to the "tenderness" effect. In other words, a material of higher damping capacity is less sensitive to the influence of surface notches or sudden changes in cross-section, which strongly affect the fatigue strength of dynamically brittle materials. Materials having high damping values are able to be deformed to a higher degree than Hooke's law predicts without being damaged. In this way stress raisers may be compensated for by internal yielding,

the amount of which is measured by the damping capacity. . . . It also seems probable that plastic action, causing the damping effect, may be related to the creep phenomenon.

Moore,[476] as stated on page 401, has termed the property by which a material can deform slightly at a locus of high stress, as at the base of a notch or crack, and can thus distribute the stress and decrease its peak value, as crackless plasticity. Such a material is resistant to the propagation of a notch. Moore pointed out the relation between crackless plasticity and damping capacity and commented that very pure metals and very fine-grained metals and alloys are most sensitive to the effect of notches. The high strength obtainable in heat-treated alloy steels, or in high-carbon steels like spring steels, may be obtained at such a sacrifice in damping power and crackless plasticity as to make them less useful than weaker carbon steels for service under concentrated or repeated stresses in notched parts.

The variations in elastic properties and in local plasticity are still too poorly understood to enable one to state their exact relationship to composition and structure. When they are better understood, it may become evident that carbon steels properly made and treated to develop the desired degree of these properties will serve for some purposes for which alloy steels are now chosen.

D. DEEP-DRAWING PROPERTIES

The ability to withstand local deformation and the tendency toward work hardening upon deformation are important factors in the evaluation of materials for deep drawing, stamping, pressing, and bending. Beside many smaller articles, automobile body and fender stock requires good deep-drawing properties. The relation between the structure of carbon steels and deep-drawing properties is discussed by Epstein in Volume I, Chapter VIII, of this monograph.

253. Requirements of a Steel for Deep Drawing.—In a steel for deep drawing it is usually desired not only that the material withstand severe deformation without too frequent anneals, but also that the surface remain smooth, without "stretcher straining" or production of an "orange peel" surface (see Volume I,

pages 274, 276, and 393). Deep-drawn objects seldom require high strength so that the properties which control formability are usually much more important than those of the finished piece after forming. Excepting stainless steels, practically all deep-drawn objects are galvanized, tinned, painted, lacquered, or enameled and are made of plain low-carbon basic open-hearth rimmed steel. The use of specially killed steels to avoid aging has been discussed in Volume I, Chapter XII. Ingot iron of about 0.02 per cent carbon is widely used, especially where the drawn objects are to be vitreous enameled; on the other hand, the carbon may sometimes be as high as 0.15 per cent, but about 0.05 to 0.10 per cent is probably the most common composition. The steel is melted and processed with great care, and much attention is paid to the control of grain size by careful box annealing or, recently and now very commonly, by normalizing, as is discussed by Epstein in Volume I, Chapter IX, of this monograph.

The evaluation of the suitability of deep-drawing sheet or strip for a given forming operation is a difficult matter, and it is not always feasible to establish the properties of a given lot of material in the laboratory and be certain that it can be formed readily into a given object or a given die without undue breakage. The user desires to get along with as few intermediate annealing operations as possible and hence arranges his practice for progressive deformation in different dies, clearance between punch and die, hold-down pressure, etc., so as practically to exhaust the ductility of good material before annealing; material which falls a little short of that ductility may show excessive breakage. The variables in the drawing process may be so different in two plants making identical objects that steel which is entirely satisfactory in one plant may be unsatisfactory in another.

254. Cupping Tests.—Many efforts have been made to evaluate the properties required of steel for deep drawing. The type of ductility needed, ability for large local deformation, would probably be indicated to some extent by reduction of area in a static tensile test, but this is very difficult to determine with accuracy on thin sheet. The tensile test does give useful information on the location of the yield strength and the rate of hardening under cold work, and also shows variations in properties in the

longitudinal and transverse directions. The tests chiefly relied upon are Rockwell-hardness testing, metallographic examination for grain size, and an Erichsen,* Olsen,* or Wazau cupping test in which a ball is pressed into the sheet which is held at the circumference. The depth of cup, and sometimes the pressure, when fracture occurs, are measured. Other shapes of plungers, or even hydraulic pressure, are sometimes used. Such tests are thought to simulate to a fair degree the deformation in deep drawing, but the correlation is far from exact. One important feature of the test is the observation of the smoothness of the surface after deformation. The Erichsen test is quite generally used in Germany for evaluation of deep-drawing properties, as is evidenced by articles by Marke[320] and by Pomp and Walther,[332] but work is under way to devise a better test. The Olsen test is used most frequently in the United States.

Sachs[762] recently modified the Erichsen test by the use of a cone-shaped plunger, in order to produce deformation more analogous to the reduction of area in a tensile test, instead of the standard plunger, which is thought to evaluate a type of ductility more analogous to elongation in the tensile test.

Siebel and Pomp[341,342,413] suggested the use of a test specimen with a hole in it; a flat punch comes down over a larger area than the hole, punching out a cup and enlarging the hole till fracture occurs somewhere in the circumference of the hole.

255. Other Tests for Deep-drawing Properties.—Sachs[485] proposed to draw a specimen cut in a wedge shape through a die of the same shape so that the upper part of the wedge, supported on all sides by the die, must either pull through the die or break off. The length of the wedge which will just pull through is taken as a criterion of suitability for deep drawing.

Tensile impact tests have been suggested since the drawing operation is usually a very rapid one and plastic flow is known to be affected by the time factor. Much of the literature to 1931 dealing with tests of material for deep drawing has been summarized by Gillett.[453]

Eksergian, in discussion of Gillett's correlated abstract,[453] pointed out that in deep drawing the general elongation, the localized ductility, the compression yield point, the tensile yield strength and tensile strength are all factors which must be

* Described in "Metals Handbook."[821]

considered; but the weight which should be given to these factors differs in the deformation of different objects; consequently, unless a laboratory test happens to provide the proper weighting, its correlation with performance in the shop will be poor. Before a material can be evaluated for its suitability for being drawn into a specified shape the critical factors of the particular deformation required in that drawing operation must be worked out and a test devised to include and weight properly all of these factors; otherwise, the different factors must be evaluated by a number of tests. There appears to be no one criterion for general suitability for drawing; hence, to compare accurately the "drawability" of an alloy steel with that of a carbon steel or of carbon steels with different carbon percentages, the comparison should be made on the basis of actual service performance in drawing a given part in a given die.

256. Olsen Ductility Tests on Low-carbon Deep-drawing Sheet.—The investigation of Nead, Mahlie, and Dittrich, which is discussed in detail on pages 101 to 104 and 135 to 138, also included Olsen ductility tests. For these tests, two specimens were cut from the center of each sheet. Impressions were made with a 1-in. die and the values averaged. The results were reported as the deviation, in thousandths of an inch, from the Inland Steel Company's standard, which is the amount of draw, in thousandths of an inch, for annealed sheet of gages varying from 0.01 to 0.11 in. of thickness. Some of the data are given in Table 109.

The conclusions from these results and from other data obtained by Nead and his associates are as follows:

1. There is little choice in drawing quality between strip flat rolled with finishing temperatures of 790 or 880°C. (1450 or 1615°F.). Strip finished at 730°C. (1345°F.) gave sheets of consistently poor ductility for all degrees of cold reductions. With high amounts of cold reduction, strip finished at 790°C. (1450°F.) gave the most ductile sheet.

2. The best deep-drawing properties result from box annealing with a sheet temperature of 705°C. (1300°F.). The results are especially good after high reductions by cold rolling.

3. Normalizing and box annealing at 650°C. (1200°F.) gave Olsen values consistently above standard at all reductions, with practically no loss in ductility for percentages of cold reduction in the critical straining range.

4. Annealing at 760°C. (1400°F.) gave considerably poorer results for all reductions than annealing at 705°C. (1300°F.). Annealing at 595°C. (1100°F.) gave poorer results than annealing at 650°C. (1200°F.), except after high cold reductions.

5. The results of the tests on the strip finished at 730°C. (1345°F.) indicate that low finishing temperatures should be avoided on low-carbon strip to be cold rolled and annealed for deep-drawing sheet.

6. Decreases in ductility in the critical-strain range (5 to 20 per cent cold reduction) were marked, especially after annealing at 595 and 650°C. (1100 and 1200°F.). Higher annealing temperatures did not cause such a marked loss of ductility.

TABLE 109.—EFFECT OF FINISHING TEMPERATURES IN HOT WORKING AND OF ANNEALING PRACTICE ON THE OLSEN DUCTILITY (EXPRESSED AS DEVIATION IN THOUSANDTHS OF AN INCH FROM INLAND STEEL COMPANY'S STANDARD FOR ANNEALED SHEET) OF FLAT-ROLLED LOW-CARBON SHEET*

Annealing temperature		Finishing temperature 730°C. (1345°F.)				Finishing temperature 790°C. (1450°F.)				Finishing temperature 880°C. (1615°F.)			
		Amount of cold reduction† before annealing, per cent											
°C.	°F.	10	25	50	65	10	25	50	65	10	25	50	65
Cold rolled		−100	−130	....	....	−63	−110	−142	....	−80	−130		
595	1100	− 30	− 5	−18	−20	−34	+ 2	− 7	+10	−38	− 5	−20	+27
650	1200	− 10	− 4	− 8	−12	−25	+ 8	− 15	+ 5	−27	0	−12	+ 7
705	1300	− 11	+ 2	+ 2	+13	− 5	+ 25	+ 30	+50	+15	+ 15	+14	+22
760	1400	− 12	− 14	+ 5	+12	− 7	+ 15	+ 6	+35	−15	+ 8	+ 6	+37
980‡	1800‡	− 17	− 30	−25	− 3	− 5	− 2	− 10	+20	+ 8	+ 3	+ 2	+10
980§	1800§	+ 22	+ 20	+26	+20	+ 8	+ 17	+ 5	+ 2	+ 1	+ 10	+ 7	+15

* Nead and associates.

† Size of hot-rolled strip before cold rolling, 0.08 in.

‡ Normalized.

§ Normalized and box annealed at 650°C. (1200°F.).

The relation between tensile properties and ductility (measured by the Olsen test) and the effect of variations in mill practice on these properties may be established by comparing the data and conclusions given on pages 101 to 104 and 135 to 138 with those given in this section.

E. MACHINABILITY

The status of refined tests to evaluate machinability of steels is much the same as in testing deep-drawing properties. All

machining processes remove material as chips, turnings, etc., but they do so in various ways; consequently, a series of steels which fall in one order of machinability for rough turning may fall in an entirely different order for finish turning or for drilling, planing, milling, broaching, etc.

257. Criteria of Machinability.—The cutting tool must penetrate below the surface to remove metal, so that indentation hardness is the first rough criterion of machinability. A hardness of 270 Brinell is often quoted as the practical limit of machinability, but with the introduction of superior high-speed steel and tungsten-carbide tools this limit is often exceeded before recourse must be had to shaping the finished section by grinding.

After the tool has penetrated, the chip must be torn off, rather than cut off, since in general the separation of metal is in advance of the tool, the tool wedging and prying off the chip. Hence, the strength and toughness of the steel, the microscopic structure and grain size are of importance as well as the hardness. Relatively weak and brittle materials machine better than strong and tough ones. A soft "gummy" material may be prone to weld to the tool and to tear, leaving a rough-machined surface. Heyn,[(159)] who described the effect of both hardness and workability or toughness in the removal of a chip, stated that in steel for automatic machines a minimum of 200 Brinell may be specified since material of lower hardness is less readily cut.

In the annealed condition, the softer, weaker steels tend to machine more easily than stronger ones, so that, broadly speaking, the higher is the carbon the more difficult is the steel to machine. In quenched and tempered steels, the further the tempering has proceeded, the easier—in general—is the machining; a completely spheroidized material often shows the best machinability, although in some cases the retention of some unspheroidized pearlite islands, to serve as more brittle stepping stones for the advancement of the cut, gives better results than a complete ferrite matrix with only globular cementite.

The criterion of machinability may be the rate at which metal can be hogged off in a roughing cut, the energy required to remove the chip, the wear of a tool as registered by a trailer tool, the length of time a standard tool will cut without being resharpened, the temperature of the chip on the tool tip, the smoothness of the finish obtained in a fine finishing cut, or others.

258. Variables Affecting Machinability.—In any given type of machining the nature of the tool, the angle and rake of the tool, the lubricant or coolant, the speed, feed, and depth of cut are variables which are just as important as the nature of the material being cut.

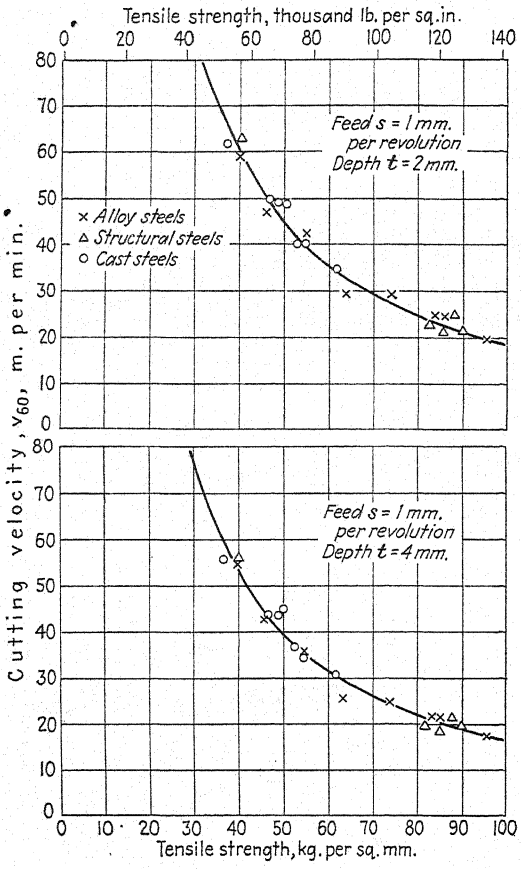

FIG. 242.—Relation of cutting speed for 1-hr. tool life to tensile strength for cuts of 2- and 4-mm. depth. (*Wallichs and Dabringhaus.*[421])

There is an extensive literature of machining and machinability. The American Society of Mechanical Engineers published a bibliography complete to 1930, compiled by Professor Boston of the University of Michigan. This has been extended recently to 1934.[352] Much of the literature refers to factors

other than the material cut. For example, Vanick and Wickenden[243] pointed out that there is a critical range of volume-removal rate within which a poor finish is obtained, but on either side of which the finish is satisfactory.

Differences in tool shape, and in speed and feed, and in type of machining operation cause large differences in the results. Boston[250] showed that no accurate conclusions on one type of machining can be drawn from results obtained on another type. Palmer[646] pointed out that for producing the same finished

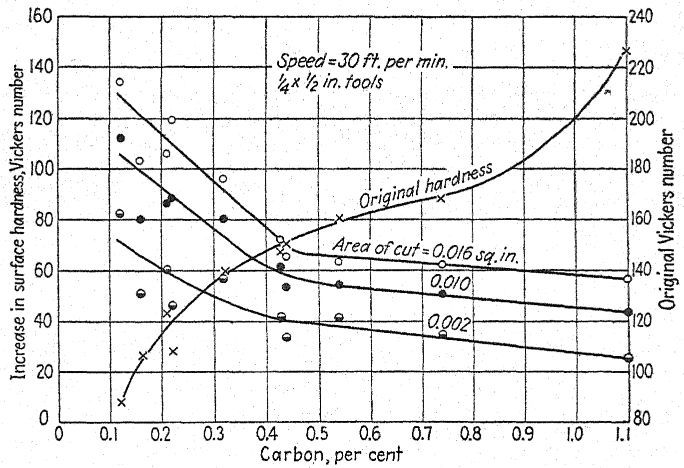

FIG. 243.—Effect of carbon content of the steel cut upon the work hardening of the surface with different areas of cuts. (*Digges.*[516])

part, one shop will want high-carbon steel in a fully spheroidized condition, another with a structure of laminated pearlite, and another with a structure in an intermediate condition. He suggested cold working the surface of a low-carbon steel, as by cold drawing, to produce some loss of toughness even at the expense of greater hardness. McMullan[637] concluded that the more readily machinable metal is one that is both soft and brittle. Bleakney[436] advocated avoidance of alumina and hard silicate inclusions, which wear the tool, and expressed preference for a pearlitic steel because it has less ductility than one fully spheroidized.

Wallichs and Dabringhaus[421] concluded that for turning, under the particular range of test conditions they used, the tensile strength or hardness is the main variable. Figure 242,

TABLE 110.—HEAT TREATMENT AND AVERAGE MECHANICAL PROPERTIES OF 0.42 PER CENT CARBON, 0.65 PER CENT MANGANESE (1040) STEEL FORGINGS CUT IN DIGGES'(447) LATHE TESTS

Forging number*	Tempering*			Tensile strength, lb. per sq. in.	Yield strength, lb. per sq. in.	Proportional limit, lb. per sq. in.	Elongation in 2 in., per cent	Reduction of area, per cent	Rockwell hardness		Shore hardness	Brinell hardness
	Temperature		Time, hr.						B	C		
	°C.	°F.										
51	260	500	6.5	123,600	85,700	69,000	17.8	55.5	100.5	20.6	37	241
51*A*	425	800	4	111,800	76,000	64,000	20.5	57.3	96.0	15.2	36	229
51*B*	595	1100	4	95,700	60,500	52,500	31.3	64.2	90.0	9.5	25	184
51*C*	705	1300	4	87,500	54,000	47,000	31.5	57.8	87.0	....	27	167
51*D*	845	1550	4	78,000	48,500	47,500	34.0	53.5	81.5	....	26	149

* The forgings were approximately 8 in. in diameter and were hollow bored with holes 3 in. in diameter. They were annealed by heating to 760 to 790°C. (1400 to 1450°F.), cooled in the furnace to 595 to 650°C. (1100 to 1200°F.), and then cooled in air to room temperature. For hardening, the forgings were heated with the furnace (3 to 4.5 hr.) to 900°C. (1650°F.), held at this temperature for 1.5 to 3.5 hr., and quenched in water. After tempering, the forgings were cooled in air with the exception of 51*D*, which was cooled slowly in the furnace.

for a series of carbon and alloy steels, is from their paper. Digges[447] included in a series of alloy steels one carbon steel of 0.42 per cent carbon, 0.19 per cent silicon, 0.65 per cent manganese, 0.02 per cent phosphorus, 0.015 per cent silicon, which he found more difficult to machine than several of the alloy steels, and to give inferior surface finish, though this was improved by partial spheroidization. The heat treatments and the machining-test data are given in Tables 110 and 111.

The carbon content has a marked effect upon the work hardening of the metal not removed, yet affected, by the cut. Figure

TABLE 111.—SUMMARY OF DIGGES'[447] LATHE TESTS* ON 0.42 PER CENT CARBON, 0.65 PER CENT MANGANESE (1040) STEEL AT 0.0115 IN. PER REVOLUTION FEED AND 0.010 IN. DEPTH OF CUT

Forging number	Tensile strength, lb. per sq. in.	Cutting speed, ft. per min.	Number of tests made	Average tool life, min.	Taylor speed,† ft. per min.
51	123,600	180	8	10.5	169
51*A*	111,800	200	8	7.4	181
51*B*	95,700	250	7	17.1	246
51*C*	87,500	310	8	19.9	310
51*D*	78,000	325	8	12.0	310

* The lathe tests were made dry with the same lot of high-speed tool steel.

† Computed from the average tool life by means of equation $VT^n = V_0T_0^n = c$, in which V = cutting speed, ft. per min.; T = tool life, min.; T_0 = 20 min.; V_0 = Taylor speed, *i.e.*, the speed which would give a tool life of 20 min.; c = a constant for the metal (with specified heat treatment) which is being cut; n = a constant experimentally determined by varying V.

243 from Digges[516] shows this. The rate at which a material work-hardens doubtless affects the machining process, a rapid rate of work hardening probably facilitating it in the case of a material which is too tough and rendering it more difficult in the case of one already near the limit of hardness.

259. Tests of Free-cutting Carbon Steels.—The effort to evaluate machinability has led to the development of many methods of testing, among them a pendulum cutter test, somewhat analogous to a pendulum impact test,[422] and many more complex tests. Such tests and simulated service, *e.g.*, setting up a screw machine for experimental use, have been especially applied to the effort of evaluating low-carbon, high-sulphur, high-phosphorus free-cutting (screw-stock) steels produced to meet

the requirements laid down by Palmer, that a free-machining material should be both soft and brittle.

Wallichs and Opitz(424) examined many free-cutting steels and concluded that no relationships can be shown between machinability and strength, chemical composition, or structure of cold-drawn screw stock. In a given series of steels, the order for ease of cutting is not the same as for good finish. Graham(704) concluded that no test method has yet evaluated the difference in machinability of different screw stocks. Boston(440) studied a series of screw stock and commented on the poor correlation between laboratory-test results and other properties of the steels, and Davis, in discussion, remarked that two screw stocks which were scarcely discernible on the basis of Professor Boston's tests nevertheless differed by 100 per cent in actual production.

Thus only a rough estimate of machinability can be made from the hardness or static properties of steels, and the effect of composition and structure, admittedly of influence, cannot yet be broadly evaluated save in terms of performance under specified machining conditions.

260. Machinability of Cast Steels.—In general the machinability of a steel casting closely approaches that of a wrought steel of the same strength and hardness. Lathe-cutting tests on a number of cast and wrought steels reported by Wallichs and Dabringhaus(421) indicated that machinability depends chiefly upon the strength and hardness of the steels tested and that cast steel has the same machinability as wrought steel of the same tensile strength.

Comprehensive studies of the machinability of cast steels were reported by Wallichs and Krekeler,(423) who used lathe tests in determining machinability, and by Wallichs and Beutel,(578) who studied the "drillability" of the steels. The materials tested, including seven carbon steels and four nickel-chromium-copper steels, contained from 0.20 to 0.35 per cent carbon and 0.52 to 1.04 per cent manganese. Silicon varied from 0.23 to 1.23 per cent. Some of the carbon steels contained almost enough sulphur and phosphorus to be classed as free cutting. The steels were made by the basic open-hearth, acid Bessemer, and basic electric processes. In the lathe tests, tools were operated to failure, and curves were drawn to show the tool life as a function of cutting speed. The cutting speed for a given depth of cut and feed that

caused the tool to fail in 60 min. was used as a measure of machinability for lathe cutting. In the drilling test, drills were operated to failure for different cutting speeds with a constant advance per revolution, and curves showing the depth drilled as a function of cutting speed were obtained. The cutting speed for a given advance per revolution that caused the drill to fail after drilling a total of 2000 mm. (78.75 in.) was used as a measure of drillability.

According to the results of Wallichs and his associates, there is no real difference in machinability between steels made by the basic open-hearth, acid Bessemer, or basic electric process.

It is well known that the skin on a casting wears cutting tools rapidly. For this reason, cast steels are considered by many as difficultly machinable. Wallichs and Krekeler[423] determined lathe-tool life for turning both the casting skin and the body of several cast steels. As would be expected, tool life in cutting the skin was short as compared with the life for turning the body of the castings.

Tests by Wallichs and Beutel[578] on two steels of approximately the same tensile strength, containing 0.22 and 0.28 per cent carbon respectively, indicated that the ease of drilling decreases as the carbon content increases. Lathe tests also showed that the lower carbon steel, even if the tensile strength was the same as that of the higher carbon material, machined more easily. The steels available to these investigators did not permit any conclusions regarding the effect of sulphur and phosphorus (in the amounts usually present in acid Bessemer steel) on machinability. The results of Wallichs and associates did show clearly, however, that, in general, by the lathe and drilling tests used, ease of machinability of cast steels decreases with increasing hardness. The tests also showed that there is more scatter in the results of drilling tests than of lathe tests.

Data on the machinability of gray cast iron, which presents a different problem from the machinability of carbon steels, are given in Chapter VIII.

F. WEAR RESISTANCE

Wear resistance is roughly inversely proportional to machinability since both involve removal of metal, resisted by hardness that makes it difficult for a tool, an abrasive particle, or a tiny

promontory on a mating metal to penetrate and get leverage for prying off a chip; thereafter it is resisted by toughness and strength of the metal, *i.e.*, the ability to resist being torn apart.

Since even under carefully controlled laboratory conditions fine gradations in machinability can often be definitely shown to exist between two materials only for one particular set of conditions, the evaluation of wear resistance in the laboratory would be expected to be accompanied by corresponding difficulties. Indeed the difficulties are even greater, for in machining, considerable volumes of metal are removed rapidly, while in wear, relatively small amounts of metal are ordinarily removed quite slowly. Duplication of the conditions producing wear in service, or accentuating them without altering the type of wear, is most difficult, and wear testing is even less standardized and less satisfactory than testing for machinability.

261. Factors Affecting Wear Resistance.—Jordan[(462)] has discussed the factors entering into different types of wear, and French[(211)] has described the types of test used in the laboratory. He showed that a test must duplicate the essential factors of service and that instead of one single measure of wear resistance there may be many measures and many orders of excellence among materials, depending upon service conditions. French designed a special wear-testing machine to simulate the wear in service on plug gages which gave results that correlated quite well with the known performance of the materials tested, but other wear tests operating under conditions dissimilar to service gave no correlation at all.

The chief differentiation in types of wear is whether the metal is in contact with non-metallic abrasives or with another metal. Two types of tests dealing with these two sets of conditions have been developed for which commercial testing equipment is on the market. In the abrasive test, proposed by Brinell, a metal disk which is not in contact with the specimen drags a standard sand or similar abrasive under light pressure across the specimen being tested. In the metal-to-metal wear test, as carried out on the Amsler equipment, two metal disks run in contact under regulated pressure, one running faster than the other so that slip occurs at the wearing surface.

262. Abrasive and Metal-to-metal Wear Resistance of Carbon Steels.—Rosenberg[(483)] summarized the results obtained on a

series of carbon steels subjected to both the abrasive and metal-to-metal types of test, as shown in Figs. 244 and 245 which are self-explanatory. He commented that in the metal-to-metal wear test free ferrite or free cementite seems detrimental to wear resistance. He found a troosto-martensitic structure to be the most resistant. In the study of wear of materials for shafts and bearings, this type of test may be run either dry or lubricated

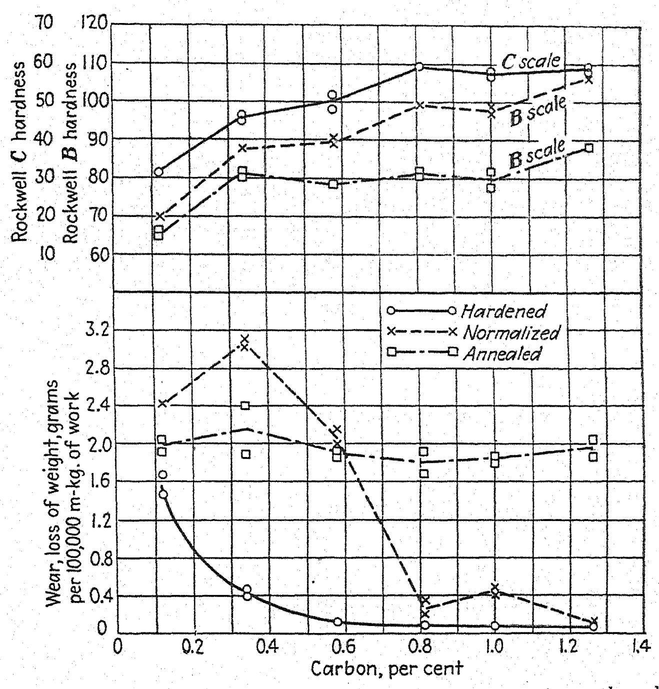

FIG. 244.—Effect of carbon content and of heat treatment on the abrasive wear of steels tested in the Brinell machine. (*Rosenberg.*[483])

A modification of such tests involves running a shaft or disk in a section of a bearing surface and noting the pressure at which seizure, *i.e.*, incipient welding of the mating surfaces, occurs.

263. Resistance of Carbon Steels to Rolling-abrasive Wear (Spindel Machine).—A rolling-abrasion testing machine largely used in Germany is the Spindel machine. Köster and Tonn[729] showed the results (Figs. 246, 247, and 248) of tests in which a steel disk 1 mm. wide and 300 to 330 mm. in diameter, of 0.40 per cent carbon steel, with a hardness of 180 Brinell, is rotated for

10 min. against the test specimen under 5 kg. pressure at 25 rev. per min. and the cut made in the specimen is measured. It

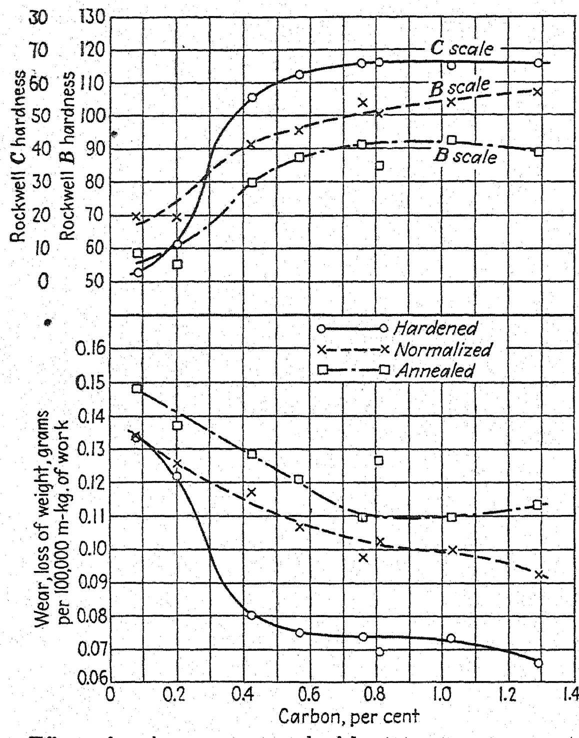

FIG. 245.—Effect of carbon content and of heat treatment on metal-to-metal wear of steels tested in the Amsler machine. (*Rosenberg.*[483])

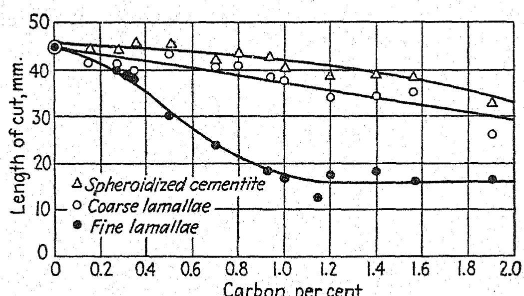

FIG. 246.—Wear of carbon steels containing spheroidized cementite, coarse lamellar pearlite, and fine lamellar pearlite. (*Köster and Tonn.*[729])

was thought that in the tests shown in Fig. 247 the soft-steel disk might not have sufficiently abraded the harder specimens,

so that a 5-mm. emery wheel was used, with results analogous to those obtained with the steel disk.

When Armco iron and a 0.28 per cent carbon steel, cold rolled with up to 60 or 70 per cent reduction, were tested by the Spindel

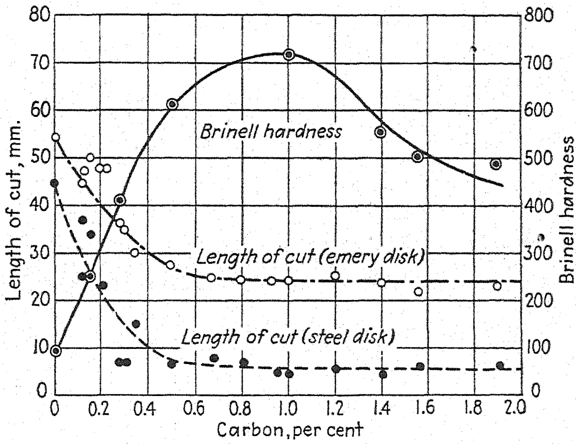

FIG. 247.—Brinell hardness and wear of carbon steels, water quenched from 1100°C. (2010°F.) (*Köster and Tonn.*[729])

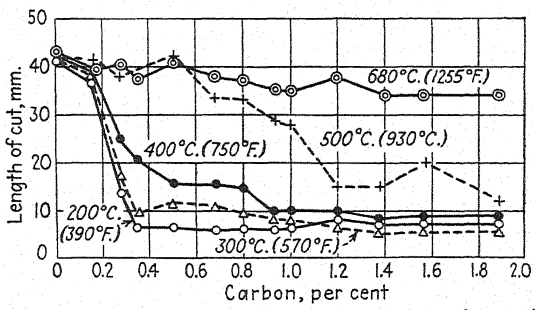

FIG. 248.—Effect of tempering on the wear of carbon steels previously water quenched from 1100°C. (2010°F.). (*Köster and Tonn.*[729])

method, the wear resistance showed no appreciable increase due to the increase in hardness obtained by cold work.

Eilender, Oertel, and Schmalz[692] made Spindel wear tests on a series of carbon steels, ranging from 0.19 to 1.75 per cent carbon, in various structural conditions, and at various speeds of rotation of the testing machine, but all under 8 kg. pressure and for 200 m. of travel against a 0.23 per cent carbon-steel disk. The maximum rate of wear in these tests, without regard to the

speed at which this maximum occurred, is plotted in Fig. 249 against Brinell hardness. The structural conditions indicated in Fig. 249 were obtained by the following heat treatments:

Lamellar pearlite. Hypoeutectoid steels heated to a temperature above line *GOS*; hypereutectoid steels heated to 750°C. (1380°F.); all cooled in air.

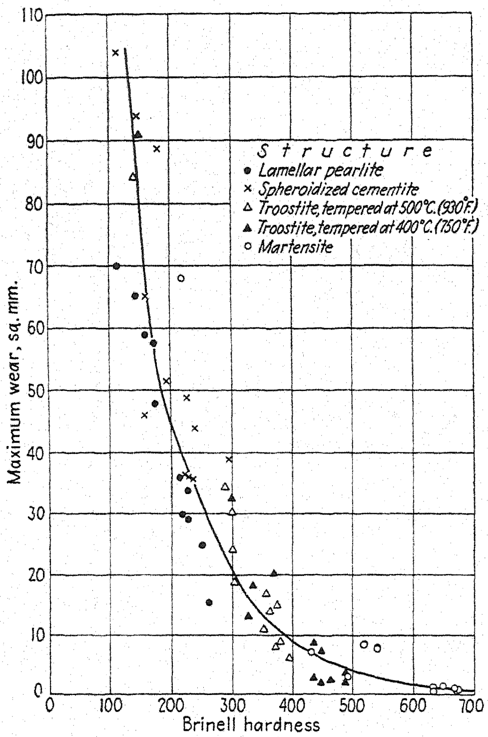

FIG. 249.—Maximum rate of wear, determined on the Spindel machine, compared with Brinell hardness of steels, containing 0.19 to 1.75 per cent carbon, in different structural conditions. (*Eilender, Oertel, and Schmalz.*[692])

Spheroidized cementite. All steels heated to a temperature above line *GOS* or at 800°C. (1470°F.) and quenched in water; reheated for 1.5 hr. by varying the temperature between 700 and 730°C. (1290 and 1345°F.) and cooled in furnace.

Troostite. All steels heated to a temperature above line *GOS* or 800°C. (1470°F.) and quenched in water; tempered 15 min. at 500 or 400°C. (930 or 750°F.).

Martensite. All steels heated as above and quenched in water; not tempered.

It was found that a structure with globular cementite was less resistant to this wear test than one with lamellar pearlite. Material with coarse grain wore less than that with fine grain. Both these findings are explained on the basis that the globules and the fine grains are more readily torn out bodily.

Several carbon steels with 1 to 1.75 per cent carbon were quenched from various temperatures up to 1100°C. (2010°F.) and tested without tempering. The wear showed no correlation with hardness or carbon content, but, when the austenite content was determined and compared, wear was found to decrease with increasing austenite up to about 15 per cent, but higher austenite contents, even up to about 50 per cent, gave about the same wear.

264. Other Data on Wear Resistance.—Tomlinson[242] made some interesting observations on the so-called rusting of steel surfaces in contact under pressure and under slight relative motion, concluding that, when seizure takes place, the motion tears off very tiny particles which are so small that they oxidize at once. The phenomenon of adhesion of clean metallic surfaces not separated by any lubricant, oxide film, or other foreign material, which may be brought about under suitable pressure and at the local temperature that can be produced by slight relative motion, probably plays a very material part in metal-to-metal wear. Methods for determination of the tendency of metals toward seizure are still in a rather chaotic state.

Fink[366] found differences in rate of wear in sliding friction depending on the atmosphere and concluded that wear was largely a phenomenon of oxidation. Investigations by Rosenberg and Jordan[760] at the National Bureau of Standards do not substantiate Fink's finding of no wear in absence of oxygen, but do show that the atmosphere has an effect.

Wear of railway rails and wheels has been discussed in the literature since 1864. Füchsel[369] included eight pages of references on this subject in a paper on rail wear.

The term "wear" is often applied to the "wearing out" of a metal part so that it fails to function, irrespective of whether actual removal of metal is involved. Thus "end batter" of rails, essentially a matter of deformation by cold working, is often considered as a type of wear. The ability of a material to work hard-

en is, of course, a factor in the process of removal of metal when high pressures are involved. Dawidenkow[362] used a repeated-blow ball-hardness tester to evaluate this property in rails, but it did not show much that was not shown by the Brinell hardness.

On the whole, the resistance of a steel to corrosion, its behavior in deep drawing, its machinability, and its wear resistance are each so fully controlled by the particular conditions of service that those conditions must be fully and exactly defined before one can compare the behavior of two steels in service. These properties are important, but the terms corrosion resistance, drawability, machinability, and wear resistance are, unfortunately, loosely used as though there were a single quantitative way of measuring them as in the evaluation of static strength by a tensile test. In reality, the properties desired are a weighted average of a whole group of separate properties. Methods of evaluation of most of these separate properties are either known or could be worked out. The difficulty occurs during the application of engineering judgment in determining which properties are of importance for the service in view so that the weighting can be accurate. Too many engineers attempt to devise some test which will differentiate between materials for a given service with little analysis of the service problem into terms of the component properties called for.

G. WELDABILITY

The process of joining metals by taking advantage of the forces of interatomic attraction which act across a joint is one of the oldest metallurgical operations. For many years forge welding of two pieces of clean iron or low-carbon steel was practiced by practically every blacksmith. Despite the age of this process, welding as a large-scale joining operation is a development of the twentieth century. The development of the various processes for commercial large-scale welding has been so rapid, has been accompanied by so much controversy, and includes so many ramifications, that a complete discussion would require a series of monographs, at least one of which could be written on the welding of carbon steels.

Faced with the problem of what to do about the subject of welding in a monograph on the properties of iron-carbon alloys, two alternatives were possible: (1) to omit all discussion to

welding, or (2) to give a summary, necessarily sketchy, of the welding of carbon steels and the variables which affect the process. The latter was adopted.

265. Welding Processes for Carbon Steels.—If clean iron or steel surfaces, heated to a temperature at which they are somewhat plastic, are brought together under pressure, they weld. This tendency to weld is utilized in the piling and rolling of wrought iron and in the making of pipe from skelp, as well as in forge or hammer welding and in the rolling of rimmed steel where the blowholes are welded. If the oxide scale on the surface is not readily fusible, it must be suitably fluxed. While rimmed steel containing less than 0.2 per cent carbon is usually preferred, medium- and high-carbon steels may be forge welded to each other or to lower carbon materials. Under suitable conditions, carbon steel can be welded to alloy steels so that "nickel-clad" and "stainless-clad" steels may be made by forge welding.

High sulphur in carbon steel for welding is considered objectionable on account of the numerous manganese sulphide inclusions; furthermore, high silicon percentages are not considered desirable, probably because of the more refractory nature of the oxide scale. Greiner, Marsh, and Stoughton,[618] however, stated that even some of the high-silicon steels can be forge welded satisfactorily.

Broadly speaking, the plain carbon steels as a class are standard for satisfactory forge welding.

In fusion welding, the edges of the pieces to be joined are melted and either pressed together or additional material of the same or another composition is melted in to fill the gap and fuse with the material to be joined. This method is widely used. Electric-resistance welding, in which the pieces to be joined are heated at the joint by a heavy current of electricity and are pressed together just as the metal at the junction fuses, plays a large part in the manufacture of automobile bodies, many automobile parts, and other objects. For this purpose, the low-carbon deep-drawing steels are suitable. Higher carbon steels, such as rails for street railways, may also be joined by resistance welding. The parts are not preheated prior to the operation in normal resistance welding.

In thermit welding, very low carbon superheated molten iron, produced by the exothermic reaction of iron oxides with metallic

aluminum, is allowed to run into a mold containing the preheated parts to be joined. This method is used for rails and for many heavy objects to which a suitable amount of molten steel cannot well be supplied by other welding methods. The smaller amounts of molten metal necessary to fill the gap in most fusion welds are supplied by carbon-arc welding, metallic-electrode welding, or by a gas (oxygen-acetylene or oxygen-hydrogen) torch. The types of welding operations are shown in Fig. 250 taken from the "Metals Handbook."(821)

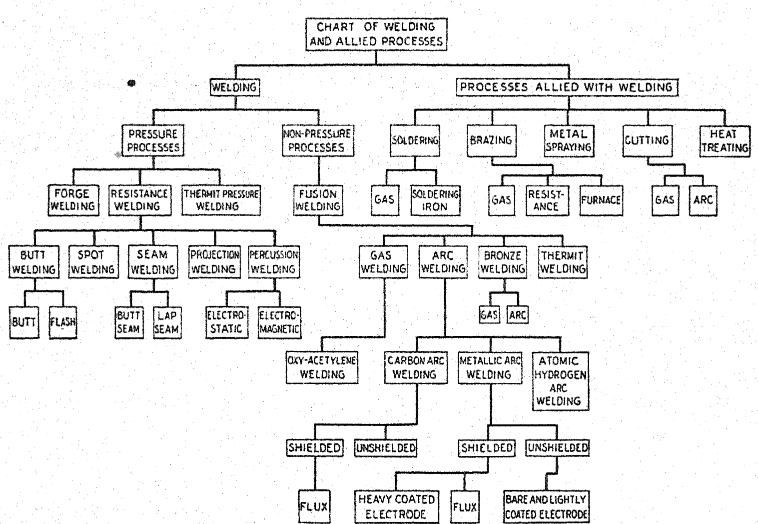

FIG. 250.—Classification of the principal welding processes. ("*Metals Handbook.*"(821))

266. General Effect of Welding on Structure.—All commercial welding processes are necessarily carried out at a high temperature. This may vary from a temperature at which the metal is fully plastic—1000 to 1350°C. (1830 to 2460°F.) depending on the carbon content—for forge welding, to one above the liquidus, for fusion welding. Obviously the use of such temperatures must have a marked effect upon the structure of the deposited metal and the sections of the parent metal adjacent to the weld.

Metal deposited in fusion welding frequently has the coarse structure characteristic of cast steels. The structure and properties of the metal in this section of the weld depend upon a

number of factors—carbon content, speed of cooling, gas and slag contents, and others. As the prime requisite of a weld is ductility and as cast high-carbon steels are generally brittle, an effort is made to keep the carbon content of the deposited metal low enough to avoid such brittleness, especially if subsequent heat treatment is impracticable. Another disadvantage of high-carbon deposited metal is that, since the cooling of the weld is frequently rapid enough to induce hardening if the carbon is high, brittleness, accompanying such hardening, may result.

In fusion and forge welding a temperature gradient exists in the parent metal over which the excessively high temperature in and adjacent to the weld decreases progressively until, at some distance, it is below the critical. As the result of this gradient all sorts of structural conditions may be present. Near the weld the grain growth may be excessive, in another section recrystallization owing to the alpha-gamma transformation may be wholly completed, and in still another section this change may be only partly completed.

In those sections where the temperature was above Ac_3 the effect of any prior heat treatment is destroyed; moreover, if the steel has been cold worked before welding, critical cold deformation may cause excessive grain growth even in sections which were not excessively overheated. All these effects may be further complicated if carbon is present in amounts necessary to make the material air hardening, or if gases which induce age hardening are present. Opposed to these deleterious effects which may result from the extreme temperature gradient is the fact that, in some cases and with some carbon percentages, cooling after welding may produce the desirable structural conditions which result from normalizing.

In addition to the various beneficial and deleterious structural conditions which result from cooling after welding, there is always the likelihood that harmful internal stresses, caused by differential heating and cooling, may be present. Thermal contraction of the weld and of the adjacent heated zones in cooling, the reversible change of some sections from alpha to gamma, and the absence of the same change in directly adjacent sections tend to set up internal strains. Subsequent heating to relieve strains to some extent is often employed; but even then, unless the material is very ductile, the weld may crack open on

cooling. The importance of avoiding internal stresses is very great. Granjon(616) pointed out that these invisible defects may be much more harmful than such visible defects as porosity.

Where the section welded is relatively small and not too irregular much of the structural heterogeneity and a large part of the internal strains can be removed by proper heat treatment. Unfortunately, however, in most commercial welding, heat treatment is impossible for reasons of design or economy.

Welds are subject to many defects if improperly made. A weld may look sound on the outside but may be incompletely fused on the inside, the added metal being merely laid down on, but not fused with, the parts being joined. The weld may contain slag or gas holes; it may pick up appreciable amounts of nitrogen from the air and thus be brittle. Many non-destructive tests have been proposed to show the quality of a weld, and X-ray radiography is now being applied on a large scale to almost all types of welds which must withstand notable stresses or whose failure would endanger the structure.* With welding becoming so important a means of fabrication, an alloy steel which cannot be readily welded is seriously handicapped and may have to give way to carbon steels of lower strength but of greater ease of welding.

267. Effect of Arc Welding on Carbon Steels.—Carbon-arc welding tends to give a higher carbon content in the weld than metallic-arc or gas welding and thus tends to produce a less ductile weld. When a weld is required which contains enough carbon to raise its strength appreciably, high-carbon electrodes or high-carbon filler rod may be used, but there is a considerable loss of carbon in the process. According to Meller(550) the loss in welding with bare electrodes of various carbon contents is as shown in the table on page 669.

When a welding rod containing about 0.10 per cent carbon and 0.40 per cent manganese is used, the weld contains about 0.05 per cent carbon and 0.10 per cent manganese when the electrode is bare. By covering the electrode with slag-forming oxides, a coating is provided which protects the weld from contact with the air and produces welds with somewhat higher carbon, much higher manganese, and much less absorption of nitrogen. Prevention of contact of the weld with the air may be brought about

* This is discussed on p. 247 in Volume I of this monograph.

Carbon Content of Electrode, Per Cent	Loss in Carbon, Per Cent
0.05	50
0.10	65
0.15	70
0.20	73
0.30	75
0.40	73
0.50	66
0.60	61
0.70	57
0.80	56
0.90	55
1.00	55

by shielding the weld with hydrogen or cracked ammonia; or the weld can be made by using a stream of superheated hydrogen (the "atomic hydrogen" process). Such welds have high ductility owing to their freedom from oxide. To test the quality of a welding rod—whether it will give a non-porous weld or not—sample welds are often laid down and examined as an acceptance test for the rod. The variables in the manufacture of the rod which are responsible for good (sound) or bad (porous) welds are discussed but seldom in the literature. Nor are there many data on the effect of composition of the base metal on the properties of the weld.

Sommer(661) prefers that the base material contain not over 0.35 per cent carbon. Leitner(630) studied the porosity of welds made with a bare electrode, containing 0.09 per cent carbon and 0.47 per cent manganese, and found that, as base material to be joined, steel practically free from silicon gave good welds, while in steels containing 0.10 to 0.44 per cent silicon the welds were porous. Increasing the silicon to 0.62 (in a steel with 0.06 per cent carbon and 0.27 per cent manganese), however, resulted in a sound weld. With carbon and silicon being about 0.10 per cent each, an increase in manganese to 1.7 per cent was required before freedom from porosity was attained. In steels of about 0.15 per cent silicon and 0.50 to 0.70 per cent manganese, the carbon had to be increased to about 0.30 per cent before the welds were sound.

Stine(777) pointed out that the base metal fuses together with the filler metal, and that the ductility of the weld must therefore depend to a considerable extent upon the composition of the

base metal. He mentioned silicon and aluminum as harmful, producing porosity, and stated that the former should be below 0.10 per cent and the latter below 0.01 per cent unless other deoxidizing elements are present. He prefers either low manganese (0.10 per cent) or high manganese (0.35 to 0.60 per cent) to amounts in the intermediate range. Carbon, according to Stine, should not be too low, 0.15 to 0.25 per cent being the best range for a general-purpose rapid-welding steel. In addition, Stine recommended that a general-purpose rapid-welding steel should contain 0.35 to 0.60 per cent manganese, a maximum of 0.07 per cent silicon, 0.05 per cent sulphur, 0.045 per cent phosphorus, and not over 2 oz. of aluminum per ton.

For a soft steel, which must withstand considerable forming, Stine recommended 0.07 to 0.15 per cent carbon, 0.35 to 0.55 per cent manganese, a maximum of 0.05 per cent silicon, and sulphur, phosphorus, and aluminum as given above. In cases where it is desired to use aluminum and silicon, the ratio of silicon to aluminum should not be less than 5 to 1 or greater than 15 to 1. Stine remarked that all commercial grades of carbon steel are weldable, but that steels of the above compositions are preferable to others for welding speed and quality of weld.

268. Effect of Welding on Mechanical Properties.—The effect of the temperature gradient adjacent to the weld on the mechanical properties of one of the most common welding steels—a 0.24 per cent carbon fire-box steel—is shown by an investigation reported by Kinzel.[805] The steel also contained 0.41 per cent manganese, a trace of silicon, 0.037 per cent sulphur, and 0.02 per cent phosphorus. Sections of 0.5-in. plate, 9 in. wide and 36 in. long, with a 45-deg. beveled edge were welded by an oxyacetylene torch. By means of holes drilled into the plate, into which the hot junctions of a number of thermocouples were inserted, it was possible to record accurately the temperature of the plate at small intervals from the scarf of the weld to a point where the welding temperature had not affected the structure and properties. The microstructure and the Rockwell *B* hardness corresponding to these temperatures could also be determined accurately. Bend tests were also made.

Other specimens of the same material were then heated in a controlled atmosphere to the same temperatures, to obtain the same microstructure and hardness which resulted from the

welding operation. According to Kinzel, "it is reasonable to suppose that the tensile properties of the specially treated samples [were] representative of the zones in the welded plate." The results obtained are given in Table 112.

Kinzel also studied the possibility of blue brittleness, which results from holding at a temperature of about 400°C. (750°F.) without an accompanying perceptible change in microstructure. Impact tests made of the plate after heating to this zone indicated that no blue brittleness would be encountered in welded sections "because the time at temperature necessary for this phenomenon to occur is of a much greater order of magnitude than that encountered in the welding operation."

The question of strain aging was also studied. Welding may set up internal strains, owing to the severe temperature gradients, unless precautions are taken to avoid them. The cold working to which many sections (pressure vessels, etc.) are subjected before welding may also introduce severe internal stresses. With cold deformations of 1.5 per cent or more, a month's aging at room temperature—or 1 hr. at 100°C. (210°F.)—reduced the impact from 40 to about 10 ft-lb. Heating 1 hr. at 550°C. (1020°F.), or 5 min. at 600°C. (1110°F.), or a normalizing treatment restored the impact to its original high value.

The conclusions regarding the effect of the temperature gradient on the properties can be stated best in Kinzel's words:

1. The heat of welding produces no noteworthy metallographic effect except in the zone immediately adjacent to the scarf. In this zone, the ductility is reduced, but the tensile strength is increased. If desired, this zone can be completely eliminated by local normalizing.

2. The minimum tensile strength in a welded joint exists in the zone subjected to the maximum subcritical temperature. This zone is not necessarily effective in reducing the strength of the joint as a whole. The high ductility of this zone, combined with the small decrease in tensile strength, permits its engineering use at the same fiber stress as that of the plate proper. This zone can be eliminated by general normalizing.

3. Commercial plate used for welding purposes is generally susceptible to strain-age brittleness. This may be produced by cold working the plate before welding, or by internal stress due to the combination of restraint and steep temperature gradients in the welding operation. This can be completely eliminated by general stress relieving.

TABLE 112.—MECHANICAL PROPERTIES OF WELDED AND MATCHED SPECIMENS OF 0.24 PER CENT CARBON PLATE*

Specimens from welded plate					Furnace-heated matched specimens								
Distance from scarf, in.	Maximum temperature reached, °C.	Maximum temperature reached, °F.	Rockwell B hardness	Bend, per cent	Temperature, °C.	Temperature, °F.	Rockwell B hardness	Bend, per cent	Tensile strength, lb. per sq. in.	Yield strength,† lb. per sq. in.	Elongation in 2 in., per cent	Reduction of area, per cent	Izod impact, ft.-lb.
1/64	1405	2560	82	....	1400	2550	82	11	81,000	46,000	14	16	8
1/32	1360	2480	80	24	1350	2460	81	28	79,000	43,000	19	31	12
1/16	1285	2345	79	....	1300	2370	79	Flat	75,000	40,000	26	49	22
1/8	1185	2165	78	Flat	1200	2190	78	Flat	71,000	40,000	32	58	28
7/32	1090	1995	76.5	Flat	1100	2010	76	Flat	69,000	40,000	35	54	28
17/32	900	1650	75	Flat	900	1650	74	Flat	70,000	40,000	35	56	28
13/16	795	1465	72.5	Flat	800	1470	74	Flat	74,000	43,000	27	52	26
1 1/8	700	1290	70	Flat	700	1290	70	Flat	64,000	36,000	38	62	52
Plate........	....	....	....	....	....	....	75	Flat	70,000	48,000	29	52	40

* Kinzel.(805)

† Average to nearest 1000 lb. per sq. in. of duplicate tests; reported as yield point.

H. AUTHOR'S SUMMARY

1. There are a number of important engineering properties of ferrous materials which cannot be measured with the preciseness that accompanies the measurement of the physical constants or the determination of such properties as tensile strength. Among these are hardness, deep-drawing characteristics, machinability, wear resistance, and weldability. Because of the uncertainty in the measurement of these qualities, and because of lack of standardization in their determination, it is almost always necessary to include in a discussion of these properties a description of the testing method used.

2. Most ferrous materials have similar strength characteristics in tension and compression. The Brinell number is a compression value, hence there is a fairly direct relation, for most carbon and low-alloy steels, between Brinell number and tensile strength. The ratio of 0.5 for tensile strength (in units of 1000 lb. per sq. in.) to Brinell hardness is an approximation of this relation and is sufficiently accurate for most engineering purposes. The actual ratio depends upon the structure and treatment, and possibly the composition, and varies from 0.45 to 0.58, with most of the determined values ranging from 0.48 to 0.54.

3. Much work has been done to establish a relationship among hardness values as determined by the different tests and between hardness and tensile strength. The data indicate that the values obtained with the Rockwell machine (one of the most flexible and widely used testers) can be converted to Brinell hardness with an expected error not greater than 10 per cent, and to tensile strength with an expected error not greater than 15 per cent. A new conversion table (Table 108, page 620) was prepared by correlating published and some unpublished data, which gives the relation among Brinell values of 197 to 514 and the corresponding Rockwell *C*, scleroscope, and Vickers-hardness values, and tensile strength. At present, because of the deformation of the ball, Brinell values are not reliable for hardness numbers above approximately 550. Owing to widely varying values in the data now available, the relation between hardness and tensile strength for Brinell values less than 197 or greater than 514 cannot be given with accuracy.

4. The single-blow notched-bar impact test, although not standardized, is widely used chiefly because materials with satisfactory impact properties in a laboratory test usually show satisfactory service in applications involving shock loading. The principal testing machines are the Izod and Charpy machines using a round or V-notch. Impact values obtained with different machines are not comparable, neither are values obtained with specimens of different sizes or with different notches. Slight differences in toughness cannot be detected by the usual impact test, but, ordinarily, brittle steels can be readily separated from tough material.

5. Cold working reduces the impact value. For specimens which have been cold worked and aged the impact is a sensitive test for susceptibility to aging. It is of value as a criterion of grain size, and to detect temper brittleness, and has been used effectively to detect microscopic internal cracks. In general, the impact resistance of carbon steel decreases as the tensile strength increases. The impact resistance is probably related roughly to the static ductility, but a variable such as grain size affects the impact values so markedly that the relation is difficult to trace. Single-blow notched-bar impact tests are of little or no value in detecting differences in such brittle materials as cast iron and hardened high-carbon steels.

6. Repeated-impact tests on notched bars give much the same type of information as the single-blow test when the intensity of the blow is high enough to break the specimen in a few thousand blows. If the intensity of the blow is so low that fracture occurs only after hundreds of thousands of blows, the information is in general the same as obtained by endurance testing. Results obtained with intermediate intensities are difficult to interpret.

7. Damping capacity is an important property of materials which must withstand severe vibration. Cast iron damps out vibration more rapidly than any other ferrous material. Carbon steel ranks next. High-strength alloy steels have, as a rule, a low damping capacity and may not, therefore, be so well suited as cast iron or carbon steel for industrial applications where severe vibration is encountered. The acoustic properties of a metal are closely related to its damping capacity; the duration of sound from a vibrating body is that portion of the damping

curve in which the vibrations are audible. Apparently there is no direct relation between damping capacity and carbon content; there are, however, data which indicate that, for a given carbon content, a steel with a pearlitic structure damps out vibration more rapidly than a steel with a troostitic structure. There may be a close relationship between damping capacity and notched-bar impact resistance, but apparently there is no relation between damping capacity and the resistance of a material to repeated stress; hence, no one has as yet been able to use the determination of damping capacity as an accelerated endurance test.

8. The ability of a material to withstand local deformation and its tendency to work harden when deformed are important characteristics of steel for deep drawing, stamping, bending, and the like. The evaluation of these characteristics by a laboratory test is difficult. The most common measure of ductility for deep drawing is the cupping test of which two forms are common: the Olsen test and the Erichsen test. The results of the cupping test are usually correlated with tensile properties and hardness but even then, unless the respective values are properly weighted, a further correlation of the laboratory results and the shop performance may be poor.

9. Data on the effect of mill practice on the deep-drawing properties of commercial low-carbon cold-rolled and annealed sheet indicate that a finishing temperature for hot rolling of 760°C. (1400°F.) or above, a reduction by cold rolling of not less than 50 per cent, and an annealing temperature of about 705°C. (1300°F.) result in the most satisfactory Olsen ductility tests. The accompanying mechanical properties (for 0.05 per cent carbon material cold rolled from 0.08-in. bands) should be 40,000 to 46,000 lb. per sq. in. tensile strength, 60 to 70 per cent yield ratio, and 40 to 50 Rockwell *B* hardness.

10. On account of the numerous variables which influence the machining characteristics of a material, many of which are more closely related to the nature and design of the tool and to coolants, speed, feed, and depth of cut than to the material cut, there is no simple relation between any mechanical property and machinability. There is, however, a rough correlation between hardness and ease of machining. Data are given which show that, for a specific machining condition, the tool life decreases as the tensile strength increases. Aside from the very general

statement that free-cutting (high-sulphur, high-phosphorus) screw stock, to be easily and rapidly machined, should be both soft and brittle, no definite relation has been established between the composition and structure of free-cutting steel and its performance in automatic machines. Summarizing, it may be said that only a rough estimate of machinability can be made from hardness and tensile properties, and that the effect of composition and structure, while undoubtedly of influence, cannot be evaluated except in terms of performance under specified conditions.

11. Wear resistance is roughly inversely proportional to machinability. Duplication of the conditions producing wear in service, or accentuating them without affecting the type of wear, is difficult; wear testing is, consequently, even less standardized and less satisfactory than testing for machinability. There seems to be no quantitative relation between hardness or carbon content and wear, although in some of the tests reported a decrease in wear was accompanied by a corresponding increase in hardness. Apparently there is a relation between wear resistance and structure; data are quoted which indicate that free ferrite and free cementite are detrimental to wear resistance, but a structure of martensite, troostite, or troosto-martensite is favorable.

12. On the whole, such properties as machinability and wear resistance (deep-drawing properties and corrosion resistance might be added as well) are so completely dependent upon the particular conditions of service that these must be completely defined and duplicated in a test before the behavior of two materials can be compared accurately. These properties are in reality the weighted average of a whole group of separate properties. Although it should be possible to evaluate each of these and weight them properly, and thus to arrive at an accurate measure of machinability and wear resistance, this has not yet been done, and no progress toward this goal will be made until engineers cease trying to devise a single test to measure the suitability of a material for a particular service when the service problem has not even been analyzed into its components.

13. Low-carbon steels are readily joined by plastic welding. Elements such as silicon or aluminum, which may produce a refractory oxide, are considered detrimental. Low-, medium-,

and high-carbon steels are readily joined by fusion welding, but to produce a sound weld precautions should be taken to prevent oxidation in and about the weld. As the deposited material usually has the coarse structure of cast metal, it is ordinarily necessary to keep the carbon in the weld metal low to prevent brittleness. The ductility of the weld depends upon the composition and structure of both parent metal and deposited metal.

14. The temperature gradient which results from fusion welding in the parent metal near the weld is accompanied by variations in structure and, frequently, high internal stresses. In welded plate containing about 0.25 per cent carbon the ductility is lower and the tensile strength higher directly next to the scarf of the weld than in the original material. The minimum tensile strength is found in the zone subjected to the highest subcritical temperature. These differences in properties, which are usually not serious, may be eliminated by simple normalizing. This treatment will also relieve stresses caused by prior cold working or by temperature gradients.

BIBLIOGRAPHY

1880–1909

1. Thurston, R. H.: Note Relating to a Peculiarity Distinguishing Annealed from Unannealed Iron, *Science*, v. 1, 1883, pp. 418–419.
2. Howe, H. M.: The Patience of Copper and Silver as Affected by Annealing, *Trans. Am. Inst. Min. Eng.*, v. 13, 1885, pp. 646–656.
3. Arnold, J. O.: The Influence of Carbon on Iron, *Minutes Proc. Inst. Civil Eng.*, v. 123, 1895, pp. 127–162.
4. Arnold, J. O.: The Properties of Steel Castings, *J. Iron Steel Inst.*, v. 59, 1901, pp. 175–204.
5. Wahlberg, A.: Brinell's Method of Determining Hardness and Other Properties of Iron and Steel, *J. Iron Steel Inst.*, v. 59, 1901, pp. 243–298; v. 60, 1901, pp. 234–271.
6. Keep, W. J.: "Cast Iron, a Record of Original Research," John Wiley & Sons, Inc., New York, 1902, 225 pp.
7. Benedicks, C.: Recherches physiques et physico-chimiques sur l'acier au carbone (Physical and Physico-chemical Investigations on Carbon Steel), Dissertation, Uppsala, 1904, 215 pp. Preliminary publ.: *Jernk. Ann.*, new ser., v. 57, 1902, pp. 124–133. Condensed: *Jernk. Ann.*, new ser., v. 61, 1906, pp. 1–231.
8. Roberts-Austen, W. C., and W. Gowland: Sixth Report to the Alloys Research Committee: On the Heat Treatment of Steel, *Proc. Inst. Mech. Eng.*, 1904, pp. 7–174.
9. Brinell, J. A.: Researches on the Comparative Hardness of Acid and Basic Open-hearth Steel at Various Temperatures by Means of "Ball Testing," *Iron Steel Magazine*, v. 9, 1905, pp. 16–19.
10. Campbell, H. H.: The Influence of Carbon, Phosphorus, Manganese, and Sulphur on the Tensile Strength of Open-hearth Steel, *Trans. Am. Inst. Min. Met. Eng.*, v. 35, 1905, pp. 772–810; discussion, pp. 1043–1048.
11. Hadfield, R. A.: Experiments Relating to the Effect on Mechanical and Other Properties of Iron and Its Alloys Produced by Liquid-air Temperatures, *J. Iron Steel Inst.*, v. 67, 1905, pp. 147–255.
12. Campbell, H. H.: "The Manufacture and Properties of Iron and Steel," McGraw-Hill Book Company, Inc., New York, 4th ed., 1907, 639 pp.
13. Meyer, E.: Untersuchungen über Härteprüfung und Härte (Hardness Testing and Hardness), *Z. Ver. deutsch. Ing.*, v. 52, 1908, pp. 645–654, 740–748, 835–844, 2077–2078.
14. Guillet, L., and L. Révillon: Expériences sur l'essai au choc à température variable (Impact Testing at Different Temperatures), *Rev. de Mét., Mém.*, v. 6, 1909, pp. 94–101.

15. KÜRTH, A.: Untersuchungen über den Einfluss der Wärme auf die Härte der Metalle (Effect of Temperature on the Hardness of Metals), *Z. Ver. deutsch. Ing.*, v. 53, 1909, pp. 85–92, 209–216.

1910–1914

16. CUSHMAN, A. S., and H. A. GARDNER: "Corrosion and Preservation of Iron and Steel," McGraw-Hill Book Company, Inc., New York, 1910, 373 pp.
17. LAMBERT, B., and J. C. THOMSON: The Wet Oxidation of Metals. 1. The Rusting of Iron, *J. Chem. Soc., Trans.*, v. 97, 1910, pp. 2426–2433.
18. ROBIN, F.: Report on the Wear of Steels and on Their Resistance to Crushing, *Iron Steel Inst., Carnegie Schol. Mem.*, v. 2, 1910, pp. 1–270.
19. FRIEND, J. A. N.: "The Corrosion of Iron and Steel," Longmans, Green & Company, New York and London, 1911, 300 pp.
20. HADFIELD, R. A., and B. HOPKINSON: The Magnetic Properties of Iron and Its Alloys in Intense Fields, *J. Inst. Elec. Eng.*, v. 46, 1911, pp. 235–306.
21. ROBIN, F.: The Variation in the Acoustic Properties of Steel with Changes of Temperature, *Iron Steel Inst., Carnegie Schol. Mem.*, v. 3, 1911, pp. 125–215.
22. BREGOWSKY, I. M., and L. W. SPRING: The Effect of High Temperature on the Physical Properties of Some Alloys, *Proc. Int. Assoc. Test. Mat.*, v. 2, May 1910–May 1913, part 2: Sixth Congress, New York, September, 1912, first section, VIII, 16 pp.
23. GUMLICH, E.: Magnetic Properties of Iron-carbon and Iron-silicon Alloys, *Trans. Faraday Soc.*, v. 8, 1912, pp. 98–114.
24. LAMBERT, B.: The Wet Oxidation of Metals. 2. The Rusting of Iron, *J. Chem. Soc., Trans.*, v. 101, 1912, pp. 2056–2057.
25. BRONIEWSKI, W.: Sur la thermo-électricité des aciers (The Thermo-electricity of Steels), *C. R. Acad. Sci.*, Paris, v. 156, 1913, pp. 1983–1985.
26. HANEMANN, H., and R. KÜHNEL: Das Verhalten gehärteter und angelassener untereutektoider Stähle (Hardened and Tempered Hypoeutectoid Steels), *Stahl u. Eisen*, v. 33, 1913, pp. 1686–1689.
27. BORNEMANN, K., and K. WAGENMANN: Die elektrische Leitfähigkeit der Metallegierungen im flüssigen Zustande (Electric Conductivity of Liquid Metallic Alloys), *Ferrum*, v. 11, 1914, pp. 276–282, 289–314, 330–343.
28. DRIESEN, J.: Untersuchungen über die thermische Ausdehnung und die Lösungsgeschwindigkeit von Kohlenstoffstählen (Thermal Expansion and the Rate of Solution of Carbon Steels), *Ferrum*, v. 11, 1914, pp. 129–138, 161–169.
29. LEA, F. C., and O. H. CROWTHER: The Change of the Modulus of Elasticity and of Other Properties of Metals with Temperature, *Engineering*, v. 98, 1914, pp. 487–489.

30. Storey, O. W.: A Study of the Annealing Process for Malleable Castings, *Met. Chem. Eng.*, v. 12, 1914, pp. 383–389.

1915–1919

31. Abbott, R. R.: The Relation between Maximum Strength, Brinell Hardness, and Scleroscope Hardness in Treated and Untreated Alloy and Plain Steels, *Proc. Am. Soc. Test. Mat.*, v. 15, part 2, 1915, pp. 42–61.
32. Burgess, G. K., and P. D. Foote: The Emissivity of Metals and Oxides. 1. Nickel Oxide (NiO) in the Range 600° to 1300°C., *Bur. Stand. Bull.* 224, v. 11, 1915, pp. 41–64.
33. Campbell, E. D.: The Influence of Heat Treatment on the Specific Resistance and Chemical Constitution of Carbon Steels, *J. Iron Steel Inst.*, v. 92, 1915, pp. 164–180.
34. Dupuy, E. L., and A. M. Portevin: The Thermoelectric Properties of Special Steels, *J. Iron Steel Inst.*, v. 91, 1915, pp. 306–335.
35. Nead, J. H.: The Effect of Carbon on the Physical Properties of Heat-treated Carbon Steel, *Trans. Am. Inst. Min. Met. Eng.*, v. 53, 1915, pp. 218–237.
36. Oberhoffer, P.: Die Bedeutung des Glühens von Stahlformguss (The Importance of Annealing Steel Castings), *Stahl u. Eisen*, v. 35, 1915, pp. 93–102, 212–214.
37. Yensen, T. D.: Magnetic and Other Properties of Iron-silicon Alloys Melted in Vacuo, Univ. of Illinois Exp. Sta., *Bull.* 83, 1915, 66 pp. Condensed: *Trans. Am. Inst. Min. Met. Eng.*, v. 53, 1915, pp. 274–305.
38. Hadfield, R. A., and J. N. Friend: Influence of Carbon and Manganese upon the Corrosion of Iron and Steel, *J. Iron Steel Inst.*, v. 93, 1916, pp. 48–76.
39. Howe, H. M.: "The Metallography of Steel and Cast Iron," McGraw-Hill Book Company, Inc., New York, 1916, 641 pp.
40. Oberhoffer, P., L. Lauber, and H. Hammel: Über die Ergebnisse von Schmiedeversuchen mit Flusseisen und Stahl (Forging Studies on Carbon Steels), *Stahl u. Eisen*, v. 36, 1916, pp. 234–238, 263–267.
41. Reinhold, O.: Über mechanische Eigenschaften von Flusseisen bei verschiedenen Temperaturen (The Mechanical Properties of Mild Steel at High Temperatures), *Ferrum*, v. 13, 1916, pp. 97–103, 116–123, 129–140.
42. Campbell, E. D.: Influence of Temperature of Measurement on the Electrical Resistivity and Thermoelectromotive Potential of Solutes in Steels, *Trans. Am. Electrochem. Soc.*, v. 32, 1917, pp. 193–214.
43. Custer, E. A.: Cast-iron Shells in Permanent Molds, *Trans. Am. Foundrymen's Assoc.*, v. 26, 1917, pp. 105–121.
44. Honda, K., and T. Simidu: On the Thermal and Electrical Conductivities of Carbon Steels at High Temperatures, *Sci. Rep.*, Sendai, ser. 1, v. 6, 1917, pp. 219–233.

45. MASON, W.: Alternating-stress Experiments, *Proc. Inst. Mech. Eng.*, 1917, pp. 121–196. Abst.: *J. Iron Steel Inst.*, v. 97, 1918, p. 537.
46. CHARPY, G.: Influence of Hot Deformation on the Qualities of Steel, *J. Iron Steel Inst.*, v. 98, 1918, pp. 7–42.
47. GUMLICH, E.: Über die Abhängigkeit der magnetischen Eigenschaften, des spezifischen Widerstandes und der Dichte der Eisenlegierungen von der chemischen Zusammensetzung und der thermischen Behandlung (Dependence of Magnetic Properties, Specific Resistance, and Density of Iron Alloys on Chemical Composition and Heat Treatment), *Wiss. Abh. Phys.-Tech. Reichsanstalt*, v. 4, 1918, pp. 267–410.
48. LAW, E. F.: Effect of Mass on Heat Treatment, *J. Iron Steel Inst.*, v. 97, 1918, pp. 333–352.
49. McWILLIAM, A.: The Influence of Some Elements on the Tenacity of Basic Steel, *J. Iron Steel Inst.*, v. 98, 1918, pp. 43–58.
50. RIDDELL, M.: Use of Iron Molds, *Foundry Tr. J.*, v. 20, 1918, pp. 195–200.
51. ELLIOTT, G. K.: Improving the Quality of Gray Iron by the Electric Furnace, *Trans. Am. Electrochem. Soc.*, v. 35, 1919, pp. 175–186.
52. HURST, J. E.: Notes on Permanent Molds and Centrifugal Castings, *Foundry Tr. J.*, v. 21, 1919, pp. 558–563.
53. WARING, F. M., and K. E. HOFAMMANN: Deep Etching of Rails and Forgings, *Proc. Am. Soc. Test. Mat.*, v. 19, part 2, 1919, pp. 182–205.

1920

54. ADAM, A. T.: Relation of Heat Treatment to Cold Work in Hypoeutectoid and Eutectoid Steels, *Iron Steel Inst., Carnegie Schol. Mem.*, v. 10, 1920, pp. 65–128.
55. British Engineering Standards Association: Report of the Steel Research Committee Founded by the Institution of Automobile Engineers and the Society of Motor Manufacturers and Traders on an Investigation Conducted to Establish the Mechanical Properties of British Standard Wrought Automobile Steels and Factors Governing Testing Procedure in Connection Therewith, *Rep.* 75, October 1920, 72 pp.
56. FOOTE, P. D., and C. O. FAIRCHILD: Optical and Radiation Pyrometry, "Symposium on Pyrometry," Am. Inst. Min. Met. Eng., New York, 1920, pp. 324–351.
57. HALL, J. H., A. E. NISSEN, and K. TAYLOR: Heat Treatment of Cast Steel, *Trans. Am. Inst. Min. Met. Eng.*, v. 62, 1920, pp. 353–396.
58. HATFIELD, W. H.: The Most Suitable Steels for Automobile Parts, *Proc. Inst. Automobile Eng.*, v. 14, 1920, pp. 503–600.
59. LANGENBERG, F. C.: Experimental Data Obtained on Charpy Impact Machine, *Trans. Am. Inst. Min. Met. Eng.*, v. 62, 1920, pp. 436–470.
60. McADAM, D. J., JR.: A High-speed Alternating Torsion Testing Machine, *Proc. Am. Soc. Test. Mat.*, v. 20, part 2, 1920, pp. 366–371.

61. Moldenke, R.: Gray-iron Castings, *Trans. Am. Soc. Mech. Eng.*, v. 42, 1920, pp. 167–174.
62. Norbury, A. L.: The Effect of Various Elements on the Electrical Resistance of Iron, *J. Iron Steel Inst.*, v. 101, 1920, pp. 627–644.
63. Schwartz, H. A.: Effect of Machining and of Cross-section on the Tensile Properties of Malleable Cast Iron, *Proc. Am. Soc. Test. Mat.*, v. 20, part 2, 1920, pp. 70–79.

1921

64. Aitchison, L.: "Engineering Steels," MacDonald and Evans, London, 1921, 396 pp.
65. Dupuy, E. L.: An Experimental Investigation of the Mechanical Properties of Steels at High Temperatures, *J. Iron Steel Inst.*, v. 104, 1921, pp. 104–116.
66. Elliott, G. K.: Superheats in the Electric Furnace, *Foundry*, v. 49, 1921, pp. 714–719.
67. French, H. J.: Effect of Temperature, Deformation, and Rate of Loading on the Tensile Properties of Low-carbon Steel below the Thermal Range, *Bur. Stand. Tech. Paper* 219, v. 16, 1921–1922, pp. 679–725.
68. Giolitti, F.: Steel Castings of High Strength and Toughness, *Chem. Met. Eng.*, v. 24, 1921, pp. 113–118, 161–165.
69. Giolitti, F.: "Heat Treatment of Soft and Medium Steels," McGraw-Hill Book Company, Inc., New York, 1921, 374 pp.
70. Hoyt, S. L.: "Metallography," Volume 2: The Metals and Common Alloys, McGraw-Hill Book Company, Inc., New York, 1921, 462 pp.
71. Jones, J. A.: Intercrystalline Cracking of Mild Steel in Salt Solutions, *Trans. Faraday Soc.*, v. 17, 1921, pp. 102–109.
72. Körber, F., and A. Dreyer: Über Blaubrüchigkeit und Altern des Eisens (Blue Brittleness and Aging of Iron), *Mitt. K.-W. Inst. Eisenforschung*, v. 2, 1921, pp. 59–90.
73. Langenberg, F. C.: Impact Properties of Various Steels, *Chem. Met. Eng.*, v. 25, 1921, pp. 910–912.
74. Langenberg, F. C.: Impact Tests on Cast Steel, *Proc. Am. Soc. Test. Mat.*, v. 21, 1921, pp. 840–851.
75. Maurer, E., and W. Schmidt: Der Einfluss verschiedener Legierungsmetalle nebst Kohlenstoff auf einige physikalische Eigenschaften des Eisens (The Influence of Various Alloying Elements Including Carbon on Some Physical Properties of Iron), *Mitt. K.-W. Inst. Eisenforschung*, v. 2, 1921, pp. 5–38.
76. Moore, H. F., and J. B. Kommers: An Investigation of the Fatigue of Metals, Univ. of Illinois Exp. Sta., *Bull.* 124, 1921, 178 pp.
77. Sykes, W. P.: Effect of Temperature, Deformation, Grain Size, and Rate of Loading on Mechanical Properties of Metals, *Trans. Am. Inst. Min. Met. Eng.*, v. 64, 1921, pp. 780–815.
78. Welter, G.: Elastizität und Festigkeit von Spezialstählen bei hohen Temperaturen (Elasticity and Strength of Special Steels at High

Temperatures), *Forschungsarbeiten a. d. Gebiete d. Ingenieurwesens*, No. 230, Verein deutscher Ingenieure, Berlin, 1921, 65 pp.

1922

79. American Society for Testing Materials: Preliminary Report of the Joint Committee on Investigation of Phosphorus and Sulphur in Steel, *Proc. Am. Soc. Test. Mat.*, v. 22, part 1, 1922, pp. 94–130.
80. Batson, R. G., and J. H. Hyde: "Mechanical Testing," E. P. Dutton & Company, Inc., New York, 1922–1923, 2 vols.
81. Bridgman, P. W.: The Compressibility of Thirty Metals as a Function of Pressure and Temperature, *Proc. Am. Acad. Arts and Sci.*, v. 58, 1922–1923, pp. 165–242.
82. Cheney, W. L.: Preparation and Properties of Pure Iron Alloys: II. Magnetic Properties of Iron-carbon Alloys as Affected by Heat Treatment and Carbon Content, *Bur. Stand. Sci. Paper* 463, v. 18, 1922–1923, pp. 609–636.
83. Elliott, G. K.: Cast Iron as Produced in the Electric Furnace and Some of Its Problems, *Trans. Am. Electrochem. Soc.*, v. 41, 1922, pp. 21–44.
84. French, H. J., and W. G. Johnson: Effect of Heat Treatment on the Mechanical Properties of One Per Cent Carbon Steel, *Trans. Am. Soc. Steel Treat.*, v. 2, 1922, pp. 467–494.
85. Guillet, L., and J. Cournot: Sur la variation des propriétés mécaniques de quelques métaux et alliages aux basses températures (Variation of Mechanical Properties of Some Metals and Alloys at Low Temperatures), *Rev. de Mét., Mém.*, v. 19, 1922, pp. 215–221.
86. Hadfield, R. A.: Corrosion of Ferrous Metals, *Min. Proc. Inst. Civil Eng.*, v. 214, 1922, pp. 83–195. Abst.: *J. Iron Steel Inst.*, v. 105, 1922, p. 644.
87. Harper, J. F., and R. S. MacPherran: Annealing Gray Cast Iron, *Trans. Am. Foundrymen's Assoc.*, v. 30, 1922, pp. 167–185.
88. Kikuta, T.: On the Growth of Gray Cast Iron during Repeated Heatings and Coolings, *Sci. Rep.*, Sendai, ser. 1, v. 11, 1922, pp. 1–17.
89. Körber, F.: Über die mechanischen Eigenschaften und das Gefüge kritisch gereckten und geglühten Weicheisens (Mechanical Properties and Structure of Critically Cold-worked and Annealed Mild Steel), *Mitt. K.-W. Inst. Eisenforschung*, v. 4, 1922, pp. 31–60.
90. Lea, F. C.: The Effect of Temperature on Some of the Properties of Metals, *Proc. Inst. Mech. Eng.*, 1922, part 2, pp. 885–912. Also: *Engineering*, v. 113, 1922, pp. 829–832.
91. Marshall, L. H.: Embrittlement of Malleable Cast Iron Resulting from Heat Treatment, *Bur. Stand. Tech. Paper* 245, v. 17, 1922–1924, pp. 677–693.
92. Phillips, A., and E. S. Davenport: Malleablizing of White Cast Iron, *Trans. Am. Inst. Min. Met. Eng.*, v. 67, 1922, pp. 466–497.
93. Rawdon, H. S., and S. Epstein: Structure of Martensitic Carbon Steels and Changes in the Microstructure Which Occur upon

Tempering, *Bur. Stand. Sci. Paper* 452, v. 18, 1922–1923, pp. 373–409.

94. SCHWARTZ, H. A.: "American Malleable Cast Iron," Penton Publishing Co., Cleveland, 1922, 416 pp.
95. WEBSTER, W. R.: Application in Rolling of Effects of Carbon, Phosphorus, and Manganese on Mechanical Properties of Steel, *Trans. Am. Inst. Min. Met. Eng.*, v. 67, 1922, pp. 220–256.
96. WÜST, F., and G. SCHITZKOWSKI: Einfluss einiger Fremdkörper auf die Schwindung des Eisens (Influence of Foreign Matter on the Shrinkage of Iron), *Mitt. K.-W. Inst. Eisenforschung*, v. 4, 1922, pp. 105–124. Condensed: *Stahl u. Eisen*, v. 43, 1923, pp. 713–720.

1923

97. BEAN, W. R.: Deterioration of Malleable in the Hot-dip Galvanizing Process, *Trans. Am. Inst. Min. Met. Eng.*, v. 69, 1923, pp. 895–931.
98. CALCOTT, W. S., J. C. WHETZEL, and H. F. WHITTAKER: "Monograph on Corrosion Tests and Materials of Construction for Chemical Engineering Apparatus," Am. Inst. Chem. Eng., New York, 1923, 182 pp.
99. FRENCH, H. J., and W. A. TUCKER: Strength of Steels at High Temperatures, *Iron Age*, v. 112, 1923, pp. 193–195, 275–278.
100. ITO, K.: The Hardness of Metals as Affected by Temperature, *Sci. Rep.*, Sendai, ser. 1, v. 12, 1923, pp. 137–148.
101. LANGENBERG, F. C.: An Investigation of the Behavior of Certain Steels under Impact at Different Temperatures, *Iron Steel Inst., Carnegie Schol. Mem.*, v. 12, 1923, pp. 75–112.
102. LANGENBERG, F. C.: Investigation of the Influence of Temperature on the Charpy Impact Value of a Group of Steels of Varying Compositions, *Yearbook Am. Iron Steel Inst.*, v. 13, 1923, pp. 349–381.
103. LEA, F. C.: The Effect of Repetition Stresses on Materials, *Engineering*, v. 115, 1923, pp. 217–219, 252–254.
104. MCADAM, D. J., JR.: Endurance Properties of Steel: Their Relation to Other Physical Properties and to Chemical Composition, *Proc. Am. Soc. Test. Mat.*, v. 23, part 2, 1923, pp. 56–129.
105. PILLING, N. B., and R. E. BEDWORTH: The Oxidation of Metals at High Temperatures, *J. Inst. Metals*, v. 29, 1923, pp. 529–591.
106. POLLITT, A. A.: "Causes and Prevention of Corrosion," Ernest Benn, Ltd., London, 1923, 230 pp.
107. REES, S. H.: Effect of Low-temperature Annealing on Some Mechanical Properties of Cold-drawn Steels, *J. Iron Steel Inst.*, v. 108, 1923, pp. 273–283.
108. WEBSTER, W. R.: Continued Discussion on the Physics of Steel, *Trans. Am. Inst. Min. Met. Eng.*, v. 69, 1923, pp. 715–721.

1924

109. AITCHISON, L., and staff: Notched Bar Impact Tests at Low Temperatures, Reports and Memoranda No. 922, Aeronautical Research

Committee, National Physical Laboratory, Engineering Dept., October, 1924, 19 pp.

110. American Society of Mechanical Engineers and American Society for Testing Materials: Symposium on Effect of Temperature upon Properties of Metals, *Proc. Am. Soc. Test. Mat.*, v. 24, part 2, 1924, pp. 9–188.

111. American Society for Testing Materials: Report of the Joint Committee on Investigation of Phosphorus and Sulphur in Steel, *Proc. Am. Soc. Test. Mat.*, v. 24, part 1, 1924, pp. 96–107.

112. Bach, C., and R. Baumann: "Elastizität und Festigkeit" (Elasticity and Strength), Julius Springer, Berlin, 9th ed., 1924, 687 pp.

113. Bancroft, W. D.: The Electrolytic Theory of Corrosion, *J. Phys. Chem.*, v. 28, 1924, pp. 785–871.

114. Benedicks, C., D. W. Berlin, and G. Phragmén: A Method for the Determination of the Specific Gravity of Liquid Iron and Other Metals of High Melting Point, *Iron Steel Inst., Carnegie Schol. Mem.*, v. 13, 1924, pp. 129–166.

115. Dowdell, R. L.: Investigation of the Treatment of Steel for Permanent Magnets, *Trans. Am. Soc. Steel Treat.*, v. 5, 1924, pp. 27–66.

116. Eicken, H., and W. Heidenhain: Einfluss von Ziehgeschwindigkeit, Querschnittsabnahme und Kohlenstoffgehalt auf die mechanischen Eigenschaften gezogener Stahldrähte. Kraftbedarf beim Ziehen (Influence of Speed of Drawing, Reduction of Cross-section, and Carbon Content on the Mechanical Properties of Drawn Steel Wires. Energy Consumption in Drawing), *Stahl u. Eisen*, v. 44, 1924, pp. 1687–1694.

117. Freeman, J. R., Jr., and A. T. Derry: Effect of Hot-rolling Conditions on the Physical Properties of a Carbon Steel, *Bur. Stand. Tech. Paper* 267, v. 18, 1924–1925, pp. 547–566.

118. Gale, R. C.: Note on "Temper" Colors, *J. Soc. Chem. Ind.*, v. 43, 1924, pp. 349–352T.

119. Gillett, H. W., and E. L. Mack: Notes on Some Endurance Tests of Metals, *Proc. Am. Soc. Test. Mat.*, v. 24, part 2, 1924, pp. 476–546; discussion, pp. 601–617.

120. Greaves, R. H.: Gun Wire: The Effect of Low-temperature Heat Treatment on the Properties of Cold-drawn Steel Wire and on Its Behavior under Stress at Raised Temperatures, Research Dept. Woolwich, *Rep.* 60, 1924, 31 pp.

121. Grotts, F.: Broadening the Field for Steel Castings through the Use of Alloys and Heat Treatment, *Trans. Am. Foundrymen's Assoc.*, v. 32, part 1, 1924, pp. 730–745.

122. Harper, J. F., and H. J. Stein: The Microscope as a Control Instrument in Annealing Steel Castings, *Trans. Am. Foundrymen's Assoc.*, v. 32, part 1, 1924, pp. 679–692.

123. Hoyt, S. L.: The Ball Indentation Hardness Test, *Trans. Am. Soc. Steel Treat.*, v. 6, 1924, pp. 396–420.

124. Jasper, T. M.: Determination of Poisson's Ratio and a Suggestion for Its Use in Stress Analysis, *Proc. Am. Soc. Test. Mat.*, v. 24, part 2, 1924, pp. 1012–1023.

125. JEFFRIES, Z., and R. S. ARCHER: "The Science of Metals," McGraw-Hill Book Company, Inc., New York, 1924, 460 pp.

126. KÖRBER, F., and A. POMP: Einfluss der Vorbehandlung auf die Kerbzähigkeit des Flusseisens in der Kälte und Wärme (Influence of Previous Treatment on the Notch Toughness of Mild Steel at High and Low Temperatures), *Mitt. K.-W. Inst. Eisenforschung*, v. 6, 1924, pp. 33–41.

127. LESSELLS, J. M.: Carbon and Carbon-vanadium Steel Castings—A Comparison, *Trans. Am. Soc. Steel Treat.*, v. 5, 1924, pp. 144–157.

128. LESTER, H. H.: X-ray Tests Applied to the Problems of the Steel Foundry, *Trans. Am. Soc. Steel Treat.*, v. 6, 1924, pp. 575–605.

129. MACPHERRAN, R. S., and J. F. HARPER: Spheroidized Cementite in Hypoeutectoid Steel, *Trans. Am. Soc. Steel Treat.*, v. 6, 1924, pp. 341–374.

130. MAURER, E.: Über ein Gusseisen Diagramm (A Cast-iron Diagram), *Kruppsche Monatshefte*, v. 5, 1924, pp. 115–122. Abst.: *Stahl u. Eisen*, v. 44, 1924, pp. 1522–1524.

131. MCADAM, D. J., JR.: Accelerated Fatigue Tests and Some Endurance Properties of Metals, *Proc. Am. Soc. Test. Mat.*, v. 24, part 2, 1924, pp. 454–475.

132. MOORE, H. F., and T. M. JASPER: An Investigation of the Fatigue of Metals, Univ. of Illinois Exp. Sta., *Bull.* 142, 1924, 86 pp.

133. O'NEILL, H.: The Effect of Cold Work upon the Density of Crystals of Alpha Iron, *J. Iron Steel Inst.*, v. 109, 1924, pp. 93–128.

134. PARDUN, C.: Über die wissenschaftlichen Grundlagen des Schleudergusses (The Scientific Principles of Centrifugal Casting), *Stahl u. Eisen*, v. 44, 1924, pp. 905–911, 1044–1048, 1200–1208. Abst.: *Foundry Tr. J.*, v. 30, 1924, pp. 295–296.

135. PETRENKO, S. N.: Physical Properties of Materials, *Bur. Stand. Circ.* 101, 2d ed., 1924, 204 pp.

136. ROTHER, W. H.: Strength of Cast Iron and Its Thickness, *Iron Age*, v. 114, 1924, pp. 326–327.

137. SAUVEUR, A.: What Is Steel?, *Trans. Am. Inst. Min. Met. Eng.*, v. 70, 1924, pp. 3–24.

138. SCHARLIBBE, L.: Discussion of paper by F. Körber: Einfluss der Temperatur auf die Festigkeitseigenschaften von Stahlguss (Effect of Temperature on Tensile Properties of Cast Steel), *Stahl u. Eisen*, v. 44, 1924, pp. 1765–1771.

139. SMITH, R. H.: High Tensile Strengths with Low-carbon Steels, *Proc. Am. Soc. Test. Mat.*, v. 24, part 2, 1924, pp. 618–625.

140. SPALDING, S. C.: Comparison of Brinell and Rockwell Hardness of Hardened High Speed Steel, *Trans. Am. Soc. Steel Treat.*, v. 6, 1924, pp. 499–504. Abst.: *American Machinist*, v. 61, 1924, pp. 759, 760.

141. SYKES, W. P.: Tensile Properties of Some Steel Wires at Liquid-air Temperatures, *Trans. Am. Soc. Steel Treat.*, v. 6, 1924, pp. 138–144.

142. WILHELM, R. B.: (Tensile Properties of Medium-carbon Steel at Temperatures between 20 and 500°C.), Discussion of "Symposium

on Effect of Temperature upon the Properties of Metals," *Proc. Am. Soc. Test. Mat.*, v. 24, part 2, 1924, pp. 142–153.

143. WOODVINE, J. G. R.: The Behavior of Case-hardened Parts under Fatigue Stresses, *Iron Steel Inst., Carnegie Schol. Mem.*, v. 13, 1924, pp. 197–237.

144. YENSEN, T. D.: Magnetic Properties of the Ternary Alloys Fe-Si-C, *Trans. Am. Inst. Elec. Eng.*, v. 43, 1924, pp. 145–175. Condensed: *J. Am. Inst. Elec. Eng.*, v. 43, 1924, pp. 558–567.

1925

145. ADAM, A. T.: "Wire Drawing and the Cold Working of Steel," Witherby, London, 1925, 212 pp.

146. AITCHISON, L., and L. W. JOHNSON: The Effect of Grain upon the Fatigue Strength of Steels, *J. Iron Steel Inst.*, v. 111, 1925, pp. 351–378.

147. American Society for Testing Materials: Report of Subcommittee V on Microhardness, *Proc. Am. Soc. Test. Mat.*, v. 25, part 1, 1925, pp. 442–443.

148. BARDENHEUER, P., and C. EBBEFELD: Beitrag zur Analyse des Schwindungsvorganges von weissem und grauem Gusseisen (The Shrinkage Process in White and Gray Cast Iron), *Mitt. K.-W. Inst. Eisenforschung*, v. 6, 1925, pp. 45–60. Also: *Giesserei*, v. 12, 1925, pp. 777–781. Also: *Stahl u. Eisen*, v. 45, 1925, pp. 1022–1027.

149. BENEDICKS, C.: Action of Hot Wall: a Factor of Fundamental Influence on the Rapid Corrosion of Water Tubes and Related to the Segregation in Hot Metals, *Trans. Am. Inst. Min. Met. Eng.*, v. 71, 1925, pp. 597–626.

150. CAMP, J. M., and C. B. FRANCIS: "The Making, Shaping and Treating of Steel," Carnegie Steel Co., Pittsburgh, 4th ed., 1925, 1142 pp.

151. COWDREY, I. H.: Relation between Rockwell and Brinell Hardness, *Trans. Am. Soc. Steel Treat.*, v. 7, 1925, pp. 244–251.

152. DONALDSON, J. W.: Low-temperature Heat Treatment of Special Cast Irons, *Foundry Tr. J.*, v. 31, 1925, pp. 517–522. Abst.: *J. Iron Steel Inst.*, v. 112, 1925, pp. 461–462.

153. FRENCH, H. J., and W. A. TUCKER: Flow in a Low-carbon Steel at Various Temperatures, *Bur. Stand. Tech. Paper* 296, v. 19, 1925, pp. 619–640.

154. GILLETT, H. W., and E. L. MACK: "Molybdenum, Cerium and Related Alloy Steels," Chemical Catalog Company, Inc., New York, 1925, 299 pp.

155. GREAVES, R. H., and J. A. JONES: The Effect of Temperature on the Behavior of Iron and Steel in the Notched-bar Impact Test, *J. Iron Steel Inst.*, v. 112, 1925, pp. 123–165. Also: *J. Inst. Metals*, v. 34, 1925, pp. 85–101.

156. HADFIELD, R. A.: I. Some Notes with Reference to the Author's Association with French Metallurgy (1889–1925) and the Important Part Played by France in the Past. II. Some Curious and Unex-

plained Facts Connected with Manganese Steel, *J. Soc. Chem. Ind.*, v. 44, 1925, pp. 1029–1044R.

157. HAYES, A., and H. E. FLANDERS: One of the Causes of Variations in Rates of Graphitization of White Cast Iron, *Trans. Am. Foundrymen's Assoc.*, v. 33, 1925, pp. 634–638.

158. HAYES, A., and G. C. SCOTT: The Catalysis of the Graphitization of White Cast Iron by the Use of Carbon Monoxide–Carbon Dioxide Mixtures when Applied under Pressure, *Trans. Am. Foundrymen's Assoc.*, v. 33, 1925, pp. 574–609.

159. HEYN, E.: "Physical Metallography," Translated and augmented by M. A. Grossmann, John Wiley & Sons, Inc., New York, 1925, 440 pp.

160. JASPER, T. M.: Typical Static and Fatigue Tests on Steel at Elevated Temperatures, *Proc. Am. Soc. Test. Mat.*, v. 25, part 2, 1925, pp. 27–52.

161. KÖRBER, F., and A. POMP: Festigkeitseigenschaften von Stahlguss bei erhöhter Temperatur (Tensile Properties of Cast Steel at Elevated Temperatures), *Mitt. K.-W. Inst. Eisenforschung*, v. 6, 1925, pp. 21–31.

162. KÖRBER, F., and A. POMP: Einfluss der Vorbehandlung auf die mechanischen Eigenschaften von Kohlenstoff- und legiertem Stahl, insbesondere die Kerbzähigkeit in der Kälte und Wärme (Influence of Pretreatment on the Mechanical Properties of Carbon and Alloy Steels, especially the Notch-toughness at High and Low Temperatures), *Mitt. K.-W. Inst. Eisenforschung*, v. 7, 1925, pp. 43–57.

163. MATHEWS, J. A.: Retained Austenite—A Contribution to the Metallurgy of Magnetism, *Trans. Am. Soc. Steel Treat.*, v. 8, 1925, pp. 565–588.

164. MCINTOSH, F. F., and W. L. COCKRELL: The Effect of Phosphorus on the Resistance of Low-carbon Steel to Repeated Alternating Stresses, Carnegie Inst. Tech., Min. Met. Investigations, *Bull.* 25, 1925, 31 pp.

165. MCNAMARA, C. J., and C. H. LORIG: A Consideration of the Annealing Operation in the Malleable Foundry, *Trans. Am. Foundrymen's Assoc.*, v. 33, 1925, pp. 624–633.

166. MEYER, H., and W. WESSELING: Die Änderung der Festigkeitseigenschaften überwiegend perlitischer Stähle durch Wärmebehandlung (Changes in the Tensile Properties of Pearlitic Steels due to Heat Treatment), *Stahl u. Eisen*, v. 45, 1925, pp. 1169–1173.

167. OBERHOFFER, P., and E. PIWOWARSKY: Über das Wachsen von Graueisen (The Growth of Gray Iron), *Stahl u. Eisen*, v. 45, 1925, pp. 1173–1178.

168. PARTRIDGE, J. H.: The Magnetic and Electrical Properties of Cast Iron, *J. Iron Steel Inst.*, v. 112, 1925, pp. 191–224.

169. PIWOWARSKY, E.: Über den Einfluss der Temperatur auf die Graphitbildung in Roh- und Gusseisen (The Effect of Temperature on Graphite Formation in Pig and Cast Irons), *Stahl u. Eisen*, v. 45, 1925, pp. 1455–1461.

170. SMITH, R. H.: Some Physical Properties of Low-carbon Steels, *Trans. Am. Soc. Steel Treat.*, v. 7, 1925, pp. 569–580.

171. WHITE, A. E.: Heat Treatment Data on Quality Steel Castings, *Trans. Am. Soc. Mech. Eng.*, v. 47, 1925, pp. 821–838. Abst.: *Mech. Eng.*, v. 48, 1926, pp. 497–500.

1926

172. American Society for Testing Materials: Report of Committee A-5 on Corrosion of Iron and Steel, *Proc. Am. Soc. Test. Mat.*, v. 26, part 1, 1926, pp. 129–142; v. 28, part 1, 1928, pp. 155–163; v. 32, part 1, 1932, pp. 105–109.

173. American Society for Testing Materials: Report of Committee E-1 on Methods of Testing. Appendices III and IV (Impact Testing), *Proc. Am. Soc. Test. Mat.*, v. 26, part 1, 1926, pp. 546–560.

174. BENEDICKS, C., H. BÄCKSTRÖM, and P. SEDERHOLM: Anomalies in Heat Conduction as Investigated in Spherical Steel Specimens with Some Determinations of Thermal (and Electrical) Conductivity in Iron and Carbon Steels, *J. Iron Steel Inst.*, v. 114, 1926, pp. 127–175.

175. BRUMFIELD, R. C.: Some Comparisons between Rockwell and Brinell Hardness, *Trans. Am. Soc. Steel Treat.*, v. 9, 1926, pp. 841–856.

176. EVANS, U. R.: "The Corrosion of Metals," Edward Arnold & Company, London, 2d ed., 1926, 259 pp.

177. FOX, E. J., and P. H. WILSON: The Principles of Centrifugal Casting as Applied to the Manufacture of "Spun" Iron Pipes, *Foundry Tr. J.*, v. 33, 1926, pp. 23–26, 43–48.

178. GOUGH, H. J.: "The Fatigue of Metals," Ernest Benn, Ltd., London, 1926, 304 pp.

179. GREAVES, R. H., and J. A. JONES: The Ratio of the Tensile Strength of Steel to the Brinell Hardness Number, *J. Iron Steel Inst.*, v. 113, 1926, pp. 335–353.

180. HONDA, K., and T. TANAKA: On the Moduli of Elasticity and Rigidity and Their Change Caused by Magnetisation in Different Kinds of Steels, *Sci. Rep.*, Sendai, ser. 1, v. 15, 1926, pp. 1–37.

181. HURST, J. E.: Some Notes on the Development of the Centrifugal Casting Process in Great Britain and Europe, *Trans. Am. Foundrymen's Assoc.*, v. 34, 1926, pp. 163–176.

182. IRRESBERGER, C.: Agitation Improves Iron, *Foundry*, v. 54, 1926, pp. 773–774.

183. ISHIGAKI, T.: The Change in the Hardness and Density in Iron and Steel Caused by Cold Working, *Sci. Rep.*, Sendai, ser. 1, v. 15, 1926, pp. 777–794.

184. KIKUTA, T.: On Malleable Cast Iron and the Mechanism of Its Graphitization, *Sci. Rep.*, Sendai, ser. 1, v. 15, 1926, pp. 115–156.

185. KIMBER, H. P., and S. M. UDALE: Some Considerations of Metal for Heavy Iron Castings in Permanent Molds, *Trans. Am. Foundrymen's Assoc.*, v. 34, 1926, pp. 195–203.

186. LEHMANN, O. H.: Die Abnutzung des Gusseisens bei gleitender Reibung (The Wear of Cast Iron under Sliding Friction), *Giesserei Zeitung*, v. 23, 1926, pp. 597–600, 623–627, 654–656. Abst.: *Foundry*, v. 55, 1927, pp. 722–723.

187. LUDWIK, P.: Streckgrenze, Kalt- und Warmsprödigkeit (Yield Strength, Cold and Hot Brittleness), *Z. Ver. deutsch. Ing.*, v. 70, 1926, pp. 379–386.

188. MARSHALL, L. H.: Process for Prevention of Embrittlement in Malleable Cast Iron, Am. Inst. Min. Met. Eng., *Preprint* 1572C, February, 1926, 8 pp.

189. MCADAM, D. J., JR.: Stress-strain Cycle Relationship and Corrosion Fatigue of Metals, *Proc. Am. Soc. Test. Mat.*, v. 26, part 2, 1926, pp. 224–254; discussion, pp. 269–280.

190. MOLDENKE, R.: Permanent and Long Life Molds, *Trans. Am. Foundrymen's Assoc.*, v. 34, 1926, pp. 204–213.

191. MOORE, H. F.: Tests of the Fatigue Strength of Cast Steel, Univ. of Illinois Exp. Sta., *Bull.* 156, 1926, 18 pp.

192. O'NEILL, H.: Hardness and Its Relation to the Cold-working and Machining Properties of Metals, part 1., *Iron Steel Inst., Carnegie Schol. Mem.*, v. 15, 1926, pp. 233–279.

193. PARR, S. W., and F. G. STRAUB: The Cause and Prevention of Embrittlement of Boiler Plate, *Proc. Am. Soc. Test. Mat.*, v. 26, part 2, 1926, pp. 52–91.

194. PIWOWARSKY, E.: Progress in the Production of High-test Cast Iron, *Trans. Am. Foundrymen's Assoc.*, v. 34, 1926, pp. 914–985.

195. ROTHER, W. H., and V. MAZURIE: The Strength of Cast Iron in Relation to Its Thickness, *Trans. Am. Foundrymen's Assoc.*, v. 34, 1926, pp. 746–765.

196. SCHWARTZ, H. A.: A Permanent Mold Process, *Trans. Am. Foundrymen's Assoc.*, v. 34, 1926, pp. 182–203.

197. SCHWARTZ, H. A.: Graphitization at Constant Temperature, *Trans. Am. Soc. Steel Treat.*, v. 9, 1926, pp. 883–906.

198. SCHWARTZ, H. A., and H. H. JOHNSON: Graphitization at Constant Temperature below the Critical Point, *Trans. Am. Soc. Steel Treat.*, v. 10, 1926, pp. 965–970.

199. SCOTT, H.: Dimensional Changes Accompanying the Phenomena of Tempering and Aging Tool Steels, *Trans. Am. Soc. Steel Treat.*, v. 9, 1926, pp. 277–304.

200. SPELLER, F. N.: "Corrosion, Causes and Prevention," McGraw-Hill Book Company, Inc., New York, 1926, 621 pp.

201. TALBOT, A. N., and F. E. RICHART: A Study of the Relation between Properties of Cast-iron Pipe Tested under Impact, Internal Pressure, and Flexure, and the Corresponding Properties Found in Several Kinds of Test Specimens Taken therefrom, *Proc. Am. Soc. Test. Mat.*, v. 26, part 2, 1926, pp. 185–217.

202. TURNER, T. H., and N. F. BUDGEN: "Metal Spraying," Griffin and Co., Ltd., London, 1926, 175 pp.

203. Wedemeyer, O.: Der Einfluss einer längeren Erhitzung auf die Auskristallisation von gebundenem Kohlenstoff im Gusseisen (The Influence of Prolonged Heating on the Formation of Combined Carbon in Cast Iron), *Stahl u. Eisen*, v. 46, 1926, pp. 557–560. Abst.: *Foundry Tr. J.*, v. 33, 1926, p. 411.

1927

204. American Society of Mechanical Engineers and American Society for Testing Materials: Report of Joint Research Committee on Effect of Temperature on the Properties of Metals, *Proc. Am. Soc. Test. Mat.*, v. 27, part 1, 1927, pp. 139–162. Subsequent reports, *ibid.*, v. 28, part 1, 1928, pp. 103–104; v. 30, part 1, 1930, pp. 93–95; v. 31, part 1, 1931, pp. 106–128; v. 32, part 1, 1932, pp. 148–192; v. 33, part 1, 1933, pp. 209–225; v. 34, part 1, 1934, pp. 207–216.
205. Benedicks, C., and H. Löfquist: Theory of the Growth of Cast Iron Repeatedly Heated, *J. Iron Steel Inst.*, v. 115, 1927, pp. 603–645.
206. Capron, J. D.: Centrifugal Casting Processes, *Yearbook Am. Iron Steel Inst.*, v. 17, 1927, pp. 59–78.
207. Coyle, F. B.: Properties and Heat Treatment of Cast Iron for Diesel Engines, *Trans. Am. Soc. Steel Treat.*, v. 12, 1927, pp. 446–465.
208. Cross, H. C., and E. E. Hill: Density of Hot-rolled and Heat-treated Carbon Steels, *Bur. Stand. Sci. Paper* 562, v. 22, 1927–1928, pp. 451–466.
209. Donaldson, J. W.: The Heat Treatment and Growth of Cast Iron, *Foundry Tr. J.*, v. 35, 1927, pp. 143–146, 167–171.
210. Freeman, J. R., Jr., R. L. Dowdell, and W. G. Berry: Endurance and Other Properties of Rail Steel, *Bur. Stand. Tech. Paper* 363, v. 22, 1927–1928, pp. 269–366.
211. French, H. J.: Wear Testing of Metals, *Proc. Am. Soc. Test. Mat.*, v. 27, part 2, 1927, pp. 212–234.
212. French, H. J., H. C. Cross, and A. A. Peterson: Creep in Five Steels at Different Temperatures, *Bur. Stand. Tech. Paper* 362, v. 22, 1927–1928, pp. 235–268.
213. Honda, K., and H. Endo: On the Volume Change in Cast Iron during Solidification, with a Criticism of the Double Diagram of the Iron-carbon System, *Sci. Rep.*, Sendai, ser. 1, v. 16, 1927, pp. 9–25.
214. Jones, H. A., I. Langmuir, and G. M. J. Mackay: The Rates of Evaporation and the Vapor Pressures of Tungsten, Molybdenum, Platinum, Nickel, Iron, Copper, and Silver, *Physical Rev.*, ser. 2, v. 30, 1927, pp. 201–214.
215. Lattré, G. de: Protection des métaux contre la corrosion (Protection of Metals against Corrosion), Impr. de l'édition et de l'industrie, Montrouge (Seine), 1927, 205 pp.
216. Lessells, J. M.: Fatigue Strength of Hard Steels and Their Relation to Tensile Strength, *Trans. Am. Soc. Steel Treat.*, v. 11, 1927, pp. 413–424.

217. MacKenzie, J. T.: Cast Iron Pipe Centrifugally Made in Sand Molds, *Yearbook Am. Iron Steel Inst.*, v. 17, 1927, pp. 466–480.
218. Masumoto, H.: On the Electrical and Thermal Conductivities of Carbon Steel and Cast Iron, *Sci. Rep.*, Sendai, ser. 1, v. 16, 1927, pp. 417–435.
219. Maurer, E., and P. Holtzhaussen: Das Gusseisendiagramm von Maurer bei verschiedenen Abkühlungsgeschwindigkeiten (Maurer's Cast-iron Diagram for Different Cooling Rates), *Stahl u. Eisen*, v. 47, 1927, pp. 1805–1812, 1977–1984. Abst.: *Foundry Tr. J.*, v. 37, 1927, pp. 222–224.
220. McAdam, D. J., Jr.: Corrosion Fatigue of Metals as Affected by Chemical Composition, Heat Treatment, and Cold Working, *Trans. Am. Soc. Steel Treat.*, v. 11, 1927, pp. 355–390.
221. McAdam, D. J., Jr.: Corrosion Fatigue of Non-ferrous Metals, *Proc. Am. Soc. Test. Mat.*, v. 27, part 2, 1927, pp. 102–127; discussion, pp. 134–152.
222. McAdam, D. J., Jr.: Fatigue and Corrosion Fatigue of Metals, Int. Cong. Test. Mat., Amsterdam 1927, v. 1, pp. 305–358.
223. Meyer, F.: Einwirkung einer weitgehenden Überhitzung auf Gefüge und Eigenschaften von Gusseisen (Effect of High Superheat on the Structure and Properties of Cast Iron), *Stahl u. Eisen*, v. 47, 1927, pp. 294–297.
224. Moore, H. F., and J. B. Kommers: "The Fatigue of Metals," McGraw-Hill Book Company, Inc., New York, 1927, 326 pp.
225. Moore, H. F., and S. W. Lyon: Tests of the Endurance of Gray Cast Iron under Repeated Stress, *Proc. Am. Soc. Test. Mat.*, v. 27, part 2, 1927, pp. 87–101; discussion, pp. 134–152.
226. Moore, H. F., S. W. Lyon, and N. P. Inglis: Tests of the Fatigue Strength of Cast Iron, Univ. of Illinois Exp. Sta., *Bull.* 164, 1927, 47 pp.
227. Moore, R. R.: Relationships between the Rockwell, Brinell, and Scleroscope Numbers, *Trans. Am. Soc. Steel Treat.*, v. 12, 1927, pp. 968–975.
228. Orrok, G. A.: High-pressure Steam Boilers, *Trans. Am. Soc. Mech. Eng.*, v. 49–50, 1927–1928, FSP-50-32, pp. 47–67.
229. Pomp, A., and W. Albert: Einfluss des Kaltziehens auf die Festigkeitseigenschaften und das Gefüge von nahtlosen Stahlrohren verschiedener Vorbehandlung (Influence of Cold Drawing on the Strength and Structure of Seamless Steel Tubing after Various Treatments), *Mitt. K.-W. Inst. Eisenforschung*, v. 9, 1927, pp. 53–93.
230. Pomp, A., and A. Dahmen: Entwicklung eines abgekürzten Prüfverfahrens zur Ermittlung der Dauerstandfestigkeit von Stahl bei erhöhten Temperaturen (Development of an Accelerated Creep Test Method for Steel), *Mitt. K.-W. Inst. Eisenforschung*, v. 9, 1927, pp. 33–52.
231. Priester, G. C., and F. J. Curran: Mechanical and Machining Properties of an Annealed Cast Iron, *Trans. Am. Soc. Steel Treat.*, v. 11, 1927, pp. 741–762.

232. RAWDON, H. S., and T. BERGLUND: Unusual Features in the Microstructure of Ferrite, *Bur. Stand. Sci. Paper* 571, v. 22, 1927–1928, pp. 649–718.
233. ROTH, E. L.: Effect of Heat Treatment on the Combined Carbon in Gray Cast Iron, *Trans. Am. Soc. Steel Treat.*, v. 12, 1927, pp. 27–40.
234. SAITO, D., and H. SAWAMURA: A New Phenomenon Concerning the Graphitization of White Cast Iron and Its Application to the Manufacture of Black Heart Malleable Castings, *Mem. Coll. Eng., Kyoto Imp. Univ.*, v. 5, 1927, pp. 1–59.
235. SAUERWALD, F.: Über die Volumengestaltung und den Molekularzustand des flüssigen und festen Roheisens (The Volumetric and Molecular Conditions of Liquid and Solid Pig Irons), *Stahl u. Eisen*, v. 47, 1927, pp. 27–28. Summarized from: *Z. anorg. allgem. Chem.*, v. 135, 1924, pp. 327–332; v. 149, 1925, pp. 273–282; and v. 155, 1926, pp. 1–12.
236. SHARPE, D.: The Holley Permanent Mould, *Foundry Tr. J.*, v. 35, 1927, pp. 297–298, 370.
237. SIPP, K., and F. ROLL: Das Wachsen des Gusseisens (The Growth of Cast Iron), *Giesserei Zeitung*, v. 24, 1927, pp. 229–284. Abst.: *Foundry Tr. J.*, v. 36, 1927, pp. 114–116.
238. SPOONER, T.: "Properties and Testing of Magnetic Materials," McGraw-Hill Book Company, Inc., New York, 1927, 385 pp.
239. SWINDEN, T., and G. R. BOLSOVER: Some Notes on Cold-rolled Strip Steel, *J. Iron Steel Inst.*, v. 115, 1927, pp. 569–602.
240. TAPSELL, H. J., and W. J. CLENSHAW: Properties of Materials at High Temperatures. II. Mechanical Properties of 0.51 per cent C Steel and 0.53 per cent C Cast Steel, Dept. Sci. Ind. Research, Eng. Res., *Special Rep.* 2, His Majesty's Stationery Office, London, 1927, 16 pp. Abst.: *Engineering*, v. 124, 1927, p. 837.
241. THIELMANN, H., and A. WIMMER: Über die innere Reibung von flüssigem Roheisen (Inner Friction in Liquid Pig Iron), *Stahl u. Eisen*, v. 47, 1927, pp. 389–399.
242. TOMLINSON, G. A.: The Rusting of Steel Surfaces in Contact, *Proc. Royal Soc.*, London, v. 115A, 1927, pp. 472–483.
243. VANICK, J. S., and T. H. WICKENDEN: Smooth Finish Machining of Low-carbon Plain and Alloy Steels, *Trans. Am. Soc. Steel Treat.*, v. 11, 1927, pp. 551–568.

1928

244. American Society of Mechanical Engineers: A Bibliography on Effect of Temperature on Properties of Metals, A.S.M.E. *Research Publ.*, 1928, 27 pp.
245. American Society for Testing Materials: Report of Committee A-3 on Correlation Tension and Transverse Tests of Cast Iron, *Proc. Am. Soc. Test. Mat.*, v. 28, part 1, 1928, pp. 142–145.
246. American Society for Testing Materials: Report of Committee A-5 on Corrosion of Iron and Steel, *Proc. Am. Soc. Test. Mat.*, v. 28, part 1, 1928, p. 150.

247. BARDENHEUER, P., and H. SCHMIDT: Der Einfluss der Kaltverformung und der Wärmebehandlung auf die elektrische Leitfähigkeit von Kupfer, Aluminium und Eisen (The Influence of Cold Working and of Heat Treatment on the Electric Conductivity of Copper, Aluminum, and Iron), *Mitt. K.-W. Inst. Eisenforschung*, v. 10, 1928, pp. 193–212.

248. BARDENHEUER, P., and K. L. ZEYEN: Beiträge zur Kenntnis des Graphits im grauen Gusseisen und seines Einflusses auf die Festigkeit (Graphite in Gray Iron and Its Influence on Its Strength), *Mitt. K.-W. Inst. Eisenforschung*, v. 10, 1928, pp. 23–53. Also: *Giesserei*, v. 15, 1928, pp. 354–365, 385–397, 411–420.

249. BOLTON, J. W.: On Research Problems of the Gray Iron Foundry, *Trans. Am. Foundrymen's Assoc.*, v. 36, 1928, pp. 469–512.

250. BOSTON, O. W.: Machinability of Metals, *Trans. Am. Soc. Steel Treat.*, v. 13, 1928, pp. 49–94, 154.

251. BOYLSTON, H. M.: "An Introduction to the Metallurgy of Iron and Steel," John Wiley & Sons, Inc., New York, 1928, 571 pp.

252. CURRAN, J. J., and J. H. G. WILLIAMS: Decarburization of High-carbon Steel in "Reducing" Atmospheres, *Trans. Am. Soc. Steel Treat.*, v. 14, 1928, pp. 809–830.

253. DONALDSON, J. W.: The Heat Treatment and Volume Changes of Gray Cast Irons between 15° and 600°C., *Foundry Tr. J.*, v. 39, 1928, pp. 299–303, 315–318.

254. GIRAUD, R.: Contribution à l'étude de l'étirage à froid de l'acier doux (Cold Drawing of Mild Steel), *Rev. de Mét.*, *Mém.*, v. 25, 1928, pp. 175–194, 235–246, 347–354.

255. GREENE, O. V.: Locomotive Forging Steels, *Trans. Am. Soc. Steel Treat.*, v. 13, 1928, pp. 573–602.

256. HALL, J. H.: Steel Castings for Severe Service, *Foundry Tr. J.*, v. 39, 1928, pp. 4–6.

257. HANKINS, G. A.: Torsional Fatigue Tests on Spring Steels, Dept. Sci. Ind. Research, Eng. Res., *Special Rep.* 9, His Majesty's Stationery Office, London, 1928, 24 pp.

258. HATFIELD, W. H.: "Cast Iron in the Light of Recent Research," Griffin and Co., Ltd., London, 1928, 340 pp.

259. INOKUTY, T.: Tensile Test of Steels at High Temperatures, *Sci. Rep.*, Sendai, ser. 1, v. 17, 1928, pp. 791–816.

260. KANTER, J. J., and L. W. SPRING: "Long-time" or "Flow" Tests of Carbon Steels at Various Temperatures with Particular Reference to Stresses below the Proportional Limit, *Proc. Am. Soc. Test. Mat.*, v. 28, part 2, 1928, pp. 80–116.

261. KENYON, R. L.: Armco Ingot Iron, *Trans. Am. Soc. Steel Treat.*, v. 13, 1928, pp. 240–269, 435–472.

262. KÖRBER, F., and A. POMP: Mechanische Eigenschaften von Stahlguss bei erhöhter Temperatur (Mechanical Properties of Cast Steel at Elevated Temperatures), *Mitt. K.-W. Inst. Eisenforschung*, v. 10, 1928, pp. 91–105.

263. KÖRBER, F., and G. SCHITZKOWSKI: Beitrag zur Schwindung von Stahlguss (Shrinkage of Cast Steel), *Stahl u. Eisen*, v. 48, 1928, pp. 129–135, 172–178.

264. Kuntze, W.: Der Bruch gekerbter Zugproben (The Fracture of Notched Tensile Test Bars), *Archiv f. d. Eisenhüttenwesen*, v. 2, 1928, pp. 109–117.

265. László, F.: Härte (Hardness), "Werkstoff-Handbuch Stahl und Eisen," section C-1, Stahleisen Verlag m. b. H., Düsseldorf, 1928, 9 pp.

266. Lehmann, K.: Die Herstellung von Dauerformen für Eisenguss (Production of Permanent Molds for Cast Iron), *Giesserei*, v. 15, 1928, pp. 130–132.

267. Logan, K. H., S. P. Ewing, and C. D. Yoemans: Bureau of Standards Soil Corrosion Studies, I, *Bur. Stand. Tech. Paper* 368, v. 22, 1928, pp. 447–554.

268. Lorenz, A. W.: A Modern Plant for the Heat Treatment of Miscellaneous Steel Castings, *Trans. Am. Foundrymen's Assoc.*, v. 36, 1928, pp. 141–152.

269. McAdam, D. J., Jr.: Corrosion of Metals as Affected by Time and by Cyclic Stress, *Trans. Am. Inst. Min. Met. Eng.*, v. 78, 1928, pp. 571–615.

270. McAdam, D. J., Jr.: Some Factors Involved in Corrosion and Corrosion Fatigue of Metals, *Proc. Am. Soc. Test. Mat.*, v. 28, part 2, 1928, pp. 117–158; discussion, pp. 168–173.

271. McMullan, O. W.: Effect of Quenching Temperature Change on the Properties of Quenched Steel, *Trans. Am. Soc. Steel Treat.*, v. 14, 1928, pp. 477–501.

272. Melmoth, F. A.: Variables in Steel Foundry Practice, *Trans. Am. Foundrymen's Assoc.*, v. 36, 1928, pp. 323–358.

273. Merten, W. J.: High-temperature Treatments of Castings and Forgings as Evidenced by Core Drill Tests from Heavy Sections, *Trans. Am. Soc. Steel Treat.*, v. 13, 1928, pp. 1–28.

274. Naeser, G.: Wärmeleitfähigkeit (Thermal Conductivity), "Werkstoff-Handbuch Stahl und Eisen," section B-17, Verlag Stahleisen m.b. H., Düsseldorf, 1928, 4 pp.

275. O'Neill, H.: Hardness and Its Relation to the Cold-working and Machining Properties of Metals, part II, *Iron Steel Inst., Carnegie Schol. Mem.*, v. 17, 1928, pp. 109–156.

276. Partridge, J. H.: The Magnetic and Electrical Properties of Cast Iron, *Iron Steel Inst., Carnegie Schol. Mem.*, v. 17, 1928, pp. 157–190.

277. Pearce, J. G.: The Use and Interpretation of the Transverse Test for Cast Iron, *J. Iron Steel Inst.*, v. 118, 1928, pp. 73–108.

278. Pfeil, L. B.: The Change in Tensile Strength Due to Aging of Cold-drawn Iron and Steel, *J. Iron Steel Inst.*, v. 118, 1928, pp. 167–194.

279. Piwowarsky, E., and H. Esser: Das Wachsen von Gusseisen (Growth of Cast Iron), *Giesserei*, v. 15, 1928, pp. 1265–1270.

280. Pullin, V. E.: X-ray Examination of Steel Castings, *Proc. Inst. Br. Foundrymen*, v. 22, 1928–1929, pp. 600–606. Abst.: *Foundry Tr. J.*, v. 39, 1928, pp. 357–359; illustrations, pp. 416–417.

281. Rawdon, H. S.: "Protective Metallic Coatings," Chemical Catalog Company, Inc., New York, 1928, 277 pp.

282. STÄBLEIN, F., and K. SCHROETER: Die Bestimmung der magnetischen Sättigung von Eisencarbid (Determination of Magnetic Saturation of Iron Carbide), *Z. anorg. allgem. Chem.*, v. 174, 1928, pp. 193–215.

283. WENSEL, H. T., and W. F. ROESER: Temperature Measurements of Molten Cast Iron, *Trans. Am. Foundrymen's Assoc.*, v. 36, 1928, pp. 191–212.

1929

284. American Society for Testing Materials: Report of Research Committee on Fatigue of Metals, *Proc. Am. Soc. Test. Mat.*, v. 29, part 1, 1929, pp. 172–173; v. 30, part 1, 1930, pp. 259–310; v. 31, part 1, 1931, p. 249; and v. 32, part 1, 1932, pp. 138–147.

285. ANON.: Halbdauerform für schwere Graugussstücke (Semipermanent Molds for Heavy Gray-iron Castings), *Giesserei Zeitung*, v. 26, 1929, p. 622.

286. BARDENHEUER, P., and K. L. ZEYEN: Die Überhitzung von Gusseisen (Superheating Cast Iron), *Mitt. K.-W. Inst. Eisenforschung*, v. 11, 1929, pp. 225–235. Also: *Giesserei*, v. 16, 1929, pp. 733–746.

287. BENEDICKS, C., N. ERICSSON, and G. ERICSON: Bestimmung des spezifischen Volumens von Eisen, Nickel und Eisenlegierungen im geschmolzenen Zustand (Determination of the Specific Volume of Molten Iron, Nickel, and Iron Alloys), *Archiv f. d. Eisenhüttenwesen*, v. 3, 1929, pp. 473–486. Abst.: *Stahl u. Eisen*, v. 50, 1930, pp. 332–334.

288. BOLTON, J. W.: Classification of Gray-iron Alloys, *Proc. Am. Soc. Test. Mat.*, v. 29, part 2, 1929, pp. 67–75.

289. BORNSTEIN, H.: Machinability and Wear, *Proc. Am. Soc. Test. Mat.*, v. 29, part 2, 1929, pp. 166–168.

290. CLARK, C. L., and A. E. WHITE: Stability of Metals at Elevated Temperatures, *Trans. Am. Soc. Steel Treat.*, v. 15, 1929, pp. 670–714.

291. COYLE, F. B., and D. M. HOUSTON: High Strength Cast Iron, *Trans. Am. Foundrymen's Assoc.*, v. 37, 1929, pp. 469–484.

292. CROOK, W. J.: Proposed Control Methods for Small Heat Treatment Plants, *Trans. Am. Soc. Steel Treat.*, v. 16, 1929, pp. 415–428.

293. DEAN, R. S., R. O. DAY, and J. L. GREGG: Relation of Nitrogen to Blue Heat Phenomena in Iron and Dispersion Hardening in the System Iron-nitrogen, *Trans. Am. Inst. Min. Met. Eng.*, v. 84, 1929, pp. 446–453.

294. FETTWEIS, F.: Die Kerbschlagprobe. Entwicklung und Kritik (The Notched-bar Impact Test and Its Development. A Criticism), *Archiv f. d. Eisenhüttenwesen*, v. 2, 1929, pp. 625–674.

295. FREEMAN, J. R., JR., and H. N. SOLAKIAN: Effect of Service on the Endurance Properties of Rail Steels, *Bur. Stand. J. Research*, v. 3, 1929, pp. 205–246.

296. GILLETT, H. W.: The Need for Information on Notch Propagation, *Metals & Alloys*, v. 1, 1929, pp. 114–116.

297. GIRARDET, F.: La fonte "girée" (Swirled Cast Iron), *Bull. Soc. Ind. de l'Est*, No. 193, 1929, pp. 19–39

298. GÜNTHER, K.: "Der Einfluss von Oberflächenbeschädigungen auf die Biegungsschwingungsfestigkeit" (The Influence of Surface Imperfections on Endurance), NEM Verlag, Berlin, 1929, 68 pp.
299. HAIGH, B. P.: Chemical Action in Relation to Fatigue of Metals, *Trans. Inst. Chem. Eng.*, v. 7, 1929, pp. 29–48.
300. HAIGH, B. P.: The Relative Safeties of Mild and High-tensile Alloyed Steels under Alternating and Pulsating Stresses, *J. Soc. Chem. Ind.*, v. 46, 1929, pp. 23–30R; discussion, by J. B. Kommers, *Engineering*, v. 132, 1931, p. 249.
301. HANKINS, G. A.: A Synopsis of the Present State of Knowledge of the Hardness and Abrasion Testing of Metals with Special Reference to the Work Done during the Period 1917–1927 and a Bibliography, *Proc. Inst. Mech. Eng.*, No. 1, 1929, pp. 317–387. Abst.: *Engineer*, v. 148, 1929, pp. 34–35, 61–63, 90–92.
302. HANKINS, G. A., and G. W. FORD: The Mechanical and Metallurgical Properties of Spring Steels as Revealed by Laboratory Tests, *J. Iron Steel Inst.*, v. 119, 1929, pp. 217–253.
303. HATFIELD, W. H.: The Application of Science to the Steel Industry, *Trans. Am. Soc. Steel Treat.*, v. 15, 1929, pp. 474–502, 652–669, 817–836, 968–1026; v. 16, 1929, pp. 121–154, 278–292.
304. HELLER, A.: Conversion of Hardness Numbers, *American Machinist*, v. 70, 1929, pp. 535–539, 583–586, 633–637.
305. HEUVERS, A.: Was hat der Stahlgiesser dem Konstrukteur über Lunker- und Rissbildung zu sagen (What Can the Steel Foundryman Teach the Construction Engineer about Piping and Cracking), *Stahl u. Eisen*, v. 49, 1929, pp. 1249–1254.
306. HEYDEKAMPF, G. S. VON: "Eine Dauerbiegemaschine mit schwingendem, in der Messstrecke gleichmässig beanspruchtem Probestab" (An Endurance Testing Machine), NEM Verlag, Berlin, 1929, 40 pp.
307. HOUDREMONT, E., and R. MAILÄNDER: Dauerbiegeversuche mit Stählen (Fatigue Tests on Steels), *Stahl u. Eisen*, v. 49, 1929, pp. 833–839.
308. HURST, J. E.: "Melting Iron in the Cupola," Penton Publishing Co., Inc., Cleveland, 1929, 220 pp.
309. JENKIN, C. F., and G. D. LEHMANN: High Frequency Fatigue, *Proc. Royal Soc.*, London, v. 125A, 1929, pp. 83–119.
310. JOHNSON, J. B., and T. T. OBERG: Fatigue Resistance of Some Aluminum Alloys, *Proc. Am. Soc. Test. Mat.*, v. 29, part 2, 1929, pp. 339–352.
311. KOMMERS, J. B.: The Fatigue Properties of Cast Iron, *Proc. Am. Soc. Test. Mat.*, v. 29, part 2, 1929, pp. 100–108; discussion, pp. 163–164.
312. KOMMERS, J. B.: The Relative Safeties of Mild and High-strength Steels under Fatigue Stresses, *J. Soc. Chem. Ind.*, v. 48, 1929, pp. 1223–1227.
313. KÖSTER, W.: Der Einfluss einer Wärmebehandlung unterhalb A_1 auf die Eigenschaften des technischen Eisens (The Influence of Heat Treatment below A_1 on the Properties of Commercial Iron), *Archiv f. d. Eisenhüttenwesen*, v. 2, 1929, pp. 503–522.

314. Kröhnke, O., E. Maass, and W. Beck: "Die Korrosion unter Berücksichtigung des Materialschutzes" (Corrosion and the Protection of Materials), S. Hirzel, Leipzig, 1929, 208 pp.

315. Kühle, A.: Einfluss des Alterns und Blaubruchs auf die Dauerschlagprobe (Influence of Aging and Blue Brittleness on the Repeated Impact Test), *Mitt. Forsch.-Inst. Vereinigte Stahlwerke, A. G.*, v. 1, 1929, pp. 83–102.

316. Lea, F. C.: The Penetration of Hydrogen into Metal Cathodes and Its Effects upon the Tensile Properties of Metals and Their Resistance to Repeated Stresses; with a Note on the Effect of Non-electrolytic Baths and Nickel Plating on these Properties, *Proc. Royal Soc.*, London, v. 123A, 1929, pp. 171–185.

317. Logan, K. H.: Soil Corrosion Studies 1927–1928, *Bur. Stand. J. Research*, v. 3, 1929, pp. 275–302.

318. MacPherran, R. S.: Effect of Section and Various Compositions on Physical Properties of Cast Iron, *Proc. Am. Soc. Test. Mat.*, v. 29, part 2, 1929, pp. 76–82.

319. MacPherran, R. S.: High Test Cast Iron, *Trans. Am. Foundrymen's Assoc.*, v. 37, 1929, pp. 495–500.

320. Marke, E.: Die Bedeutung der ersten Glühung für Qualitätsfeinbleche, ihr Einfluss auf die Erichsen-Tiefung und das Gefüge (The Importance of the First Annealing of High-quality Steels, and Its Effect on the Erichsen Cup Depth and the Structure), *Archiv f. d. Eisenhüttenwesen*, v. 2, 1929, pp. 851–858.

321. Maurer, E., and K. Schroeter: Die Bestimmung des Austenitgehaltes durch Messung des magnetischen Sättigungswertes und die Vorgänge beim Anlassen gehärteter Stähle (Determination of the Austenite Content by Measuring the Magnetic Saturation, and the Reactions in Tempering Hardened Steel), *Stahl u. Eisen*, v. 49, 1929, pp. 929–940.

322. McAdam, D. J., Jr.: Corrosion of Metals as Affected by Stress, Time, and Number of Cycles, *Trans. Am. Inst. Min. Met. Eng.*, v. 83, 1929, pp. 56–110; discussion, pp. 53–55.

323. McAdam, D. J., Jr.: Corrosion of Metals under Cyclic Stress, *Proc. Am. Soc. Test. Mat.*, v. 29, part 2, 1929, pp. 250–303.

324. McAdam, D. J., Jr.: Fatigue and Corrosion Fatigue of Spring Material, *Trans. Am. Soc. Mech. Eng.*, v. 51, part 1, 1929, APM-51-5, pp. 45–58.

325. Messkin, W. S.: Der Einfluss des Kaltreckens auf die magnetischen Eigenschaften eines Kohlenstoffstahles (The Influence of Cold Drawing on the Magnetic Properties of a Carbon Steel), *Archiv f. d. Eisenhüttenwesen*, v. 3, 1929, pp. 417–425.

326. Moore, H. F.: The Fatigue of Metals—A Review of Progress from 1920–1929, *Yearbook Am. Iron Steel Inst.*, v. 19, 1929, pp. 304–336.

327. Moore, H. F.: Materials in Machine Construction, *Mech. Eng.*, v. 51, 1929, pp. 739–743.

328. Norton, F. H.: "The Creep of Steel at High Temperatures," McGraw-Hill Book Company, Inc., New York, 1929, 90 pp.

329. Pfeil, L. B.: The Oxidation of Iron and Steel at High Temperatures, *J. Iron Steel Inst.*, v. 119, 1929, pp. 501–560.

330. Piwowarsky, E.: "Hochwertiger Grauguss" (High-strength Cast Iron), Julius Springer, Berlin, 1929, 336 pp.

331. Pomp, A., and H. Poellein: Festigkeits- und Gefügeuntersuchungen an kaltgewalzten und geglühten Bandstählen verschiedener Vorbehandlung (Strength and Structure of Cold-rolled and Annealed Steel Strips Subjected to Different Previous Treatments), *Archiv f. d. Eisenhüttenwesen*, v. 3, 1929, pp. 223–231.

332. Pomp, A., and L. Walther: Der Einfluss der Stichabnahme und der Glühtemperatur auf die mechanischen Eigenschaften und das Gefüge von kaltgewalzten Feinblechen (The Influence of Pass Reduction and Annealing Temperature on Mechanical Properties and Structure of Cold-rolled Thin Sheets), *Archiv f. d. Eisenhüttenwesen*, v. 2, 1929, pp. 859–865.

333. Ralston, O. C.: Iron Oxide Reduction Equilibria: A Critique from the Standpoint of the Phase Rule and Thermodynamics, *Bur. Mines Bull.* 296, 1929, 326 pp.

334. Rawdon, H. S.: Laboratory Corrosion Tests of Mild Steel with Special Reference to Ship Plate, *Bur. Stand. J. Research*, v. 2, 1929, pp. 431–440.

335. Rother, W. H., and V. Mazurie: Correlation of Cast-iron Test Bars and Castings, *Proc. Am. Soc. Test. Mat.*, v. 29, part 2, 1929, pp. 83–86.

336. St. John, A.: X-ray Inspection of Castings, *Fuels and Furnaces*, v. 7, 1929, pp. 1371–1372, 1407–1408.

337. Satoh, S.: Metallic Electrodes for Cast-iron Arc Welding, *Trans. Am. Inst. Min. Met. Eng.*, v. 84, 1929, pp. 144–165.

338. Sawamura, H.: On the Various Methods to Favor the Graphitization in White Cast Iron, World Eng. Cong., Tokyo, 1929, *Proc.*, v. 34, pp. 307–357.

339. Schikorr, G.: Über das System Eisen–Wasser (The Iron–water System), *Z. f. Elektrochemie*, v. 35, 1929, pp. 62–65.

340. Schikorr, G.: Über die Reaktionen zwischen Eisen, seinen Hydroxyden und Wasser (The Reactions between Iron, Its Hydroxides, and Water), *Z. f. Elektrochemie*, v. 35, 1929, pp. 65–70.

341. Siebel, E., and A. Pomp: Über den Kraftverlauf beim Tiefziehen und bei der Tiefungsprüfung (The Power Consumed in Deep Drawing and in the Deep-drawing Test), *Mitt. K.-W. Inst. Eisenforschung*, v. 11, 1929, pp. 139–154.

342. Siebel, E., and A. Pomp: Ein neues Prüfverfahren für Feinbleche (A New Testing Method for Thin Sheets), *Mitt. K.-W. Inst. Eisenforschung*, v. 11, 1929, pp. 287–291. Abst.: *Metallurgist*, v. 6, 1930, pp. 6–7.

343. Smalley, O.: High Test Cast Iron, *Trans. Am. Foundrymen's Assoc.*, v. 37, 1929, pp. 485–500. Abst.: *Foundry Tr. J.*, v. 42, 1930, pp. 25–26.

344. Speller, F. N., I. B. McCorkle, and P. F. Mumma: The Influence of Corrosion Accelerators and Inhibitors on Fatigue of Ferrous Metals, *Proc. Am. Soc. Test. Mat.*, v. 29, part 2, 1929, pp. 238–249; discussion, pp. 304–313.

345. STRAUSS, J.: Metals and Alloys for Industrial Applications Requiring Extreme Stability, *Trans. Am. Soc. Steel Treat.*, v. 16, 1929, pp. 191–226.

346. STUMPER, R.: Untersuchungen über den Einfluss der Walztemperatur auf die Eigenschaften der Schienen (Influence of Rolling Temperature on the Properties of Rails), *Stahl u. Eisen*, v. 49, 1929, pp. 177–187; discussion pp. 843–845.

347. THUM, A.: Neuere Anschauungen über die mechanischen Eigenschaften des Gusseisens (New Opinions on the Mechanical Properties of Cast Iron), *Giesserei*, v. 16, 1929, pp. 1164–1174.

348. THUM, A., and H. UDE: Die Elastizität und Schwingungsfestigkeit des Gusseisens (Elasticity and Alternate Strength of Cast Iron), *Giesserei*, v. 16, 1929, p. 501.

349. WALLICHS, A., and H. SCHALLBROCH: Die Beziehung zwischen verschiedenen Härteziffern bei ungehärteten Kohlenstoffstählen (The Relation between Different Hardness Numbers in Unhardened Carbon Steels), *Maschinenbau*, v. 8, 1929, pp. 69–74. Abst.: *Stahl u. Eisen*, v. 49, 1929, p. 1349.

350. WALLIN, H. N., and F. B. OLCOTT: Corrosion Resisting Steel, Navy Dept., Bur. of Construction and Repair, *Tech. Bull.* 6, September, 1929, 26 pp.

351. ZANDER, W.: "Der Einfluss von Oberflächenbeschädigungen auf die Biegungsschwingungsfestigkeit" (Influence of Surface Defects on Fatigue), NEM Verlag, Berlin, 1929, 65 pp.

1930

352. American Society of Mechanical Engineers: "A Bibliography on the Cutting of Metals," (compiled by O. W. Boston). Part I, A. S. M. E. *Research Publ.*, 1930, 63 pp. Part II, publ. by the compiler, Ann Arbor, 1935, 202 pp.

353. American Society for Testing Materials: Report of Committee A-2 on Wrought Iron, *Proc. Am. Soc. Test. Mat.*, v. 30, part 1, 1930, pp. 159–186.

354. ANON.: Giessen kleiner Graugussmassenteile in Dauerformen (Making Small Gray-iron Castings in Permanent Molds), *Giesserei*, v. 17, 1930, pp. 534–535.

355. BARKLIE, R. H. D., and H. J. DAVIES: The Effect of Surface Conditions and Electrodeposited Metals on the Resistance of Materials to Repeated Stresses, *Proc. Inst. Mech. Eng.*, 1930, part 1, pp. 731–750. Also: *Engineer*, v. 150, 1930, pp. 670–673.

356. BAUER, O., and K. SIPP: Der Einfluss von Phosphor, Schwefel, Nickel und Chrom auf das Wachstum des Gusseisens (The Influence of Phosphorus, Sulphur, Nickel, and Chromium on the Growth of Cast Iron), *Giesserei*, v. 17, 1930, pp. 989–995.

357. BELAIEW, N. T.: Sur la perlite, sa structure et ses propriétés mécaniques (Pearlite, Its Structure and Mechanical Properties), *Rev. de Mét., Mém.*, v. 27, 1930, pp. 680–685.

358. Bierbaum, C. H.: The Microcharacter, *Trans. Am. Soc. Steel Treat.*, v. 18, 1930, pp. 1009–1025.
359. Chickaschige, M., and K. Uno: Die Beziehungen zwischen der Farbe und dem Feingefüge der Legierungen. IV. Eisen-Kohlenstoff-Legierungen (The Relation between Color and the Microstructure of Alloys. IV. Iron-carbon Alloys), *Z. anorg. allgem. Chem.*, v. 190, 1930, pp. 303–312.
360. Cowdrey, I. H.: Hardness by Mutual Indentation, *Proc. Am. Soc. Test. Mat.*, v. 30, part 2, 1930, pp. 559–571.
361. Daeves, K.: Die mechanischen Eigenschaften von Thomas- und Siemens-Martin-Stählen (The Mechanical Properties of Basic Bessemer and Acid Open-hearth Steels), *Stahl u. Eisen*, v. 50, 1930, pp. 1353–1356. Abst.: *J. Iron Steel Inst.*, v. 122, 1930, pp. 578–579.
362. Dawidenkow, N.: Die Verformungsabnutzung der Schienen (Deformation Wear of Rails), *Stahl u. Eisen*, v. 50, 1930, pp. 1580–1584.
363. Dickson, T. C.: Casting Guns by the Centrifugal Process, *Trans. Am. Soc. Steel Treat.*, v. 18, 1930, pp. 212–240.
364. Donaldson, J. W.: Thermal Conductivity of Metals and Alloys, *Metallurgia*, v. 1, 1930, pp. 119–120.
365. Döring, H.: Das Drücken der Oberfläche und der Einfluss von Querbohrungen auf die Biegungsschwingungsfestigkeit (Surface Working and Influence of Transverse Holes on Fatigue Resistance), *Veröff. d. Wöhler-Inst., Braunschweig*, No. 5, NEM Verlag, Berlin, 1930, 50 pp. Extended Abst.: *Metals & Alloys*, v. 2, 1931, pp. 94–95.
366. Fink, M.: Wear Oxidation, a New Component of Wear, *Trans. Am. Soc. Steel Treat.*, v. 18, 1930, pp. 1027–1034.
367. Freeman, J. R., Jr., and R. D. France: Endurance Properties of Some Special Rail Steels, *Bur. Stand. J. Research*, v. 4, 1930, pp. 851–874.
368. Freeman, J. R., Jr., and G. W. Quick: Tensile Properties of Rail and Some Other Steels at Elevated Temperatures, *Bur. Stand. J. Research*, v. 4, 1930, pp. 549–591. Also: *Trans. Am. Inst. Min. Met. Eng.*, v. 90, 1930, pp. 225–279.
369. Füchsel, M.: Über Verschleissbarkeit der Werkstoffe bei trockener Reibung (The Wear of Materials under Dry Friction), First Communication, New Int. Assoc. Test. Mat., Group D, 1930, pp. 95–112.
370. Gillett, H. W.: More on Fatigue and High Frequency Fatigue Testing, *Metals & Alloys*, v. 1, 1930, pp. 332–333.
371. Gough, H. J., and A. J. Murphy: The Effect of Low Temperatures on the Shock-resisting Properties of New Wrought-iron Chain, *Proc. Inst. Mech. Eng.*, No. 5, 1930, pp. 1159–1223.
372. Greaves, R. H.: Meaning of the Notched-bar Impact Test for Investigation and for Acceptance Test Purposes, First Communication, New Int. Assoc. Test. Mat., Group A, 1930, pp. 225–229.
373. Greulich, W.: Die Verfestigung einiger Werkstoffe beim Kaltwalzen (The Work Hardening of Some Materials on Cold Rolling), *Stahl*

u. Eisen, v. 50, 1930, pp. 1397–1401. Abst.: *J. Iron Steel Inst.*, v. 122, 1930, pp. 580–581.

374. HARVEY, W. E.: Zinc as a Protective Coating against Corrosion Fatigue of Steel, *Metals & Alloys*, v. 1, 1930, pp. 458–461.
375. HASE, R.: Einfluss des Emissionsvermögens auf die Temperaturmessung am flüssigen Eisen (Influence of Emissive Power on Temperature Measurements of Liquid Iron), *Archiv f. d. Eisenhüttenwesen*, v. 4, 1930, pp. 261–264.
376. HATFIELD, W. H.: Permanence of Dimensions under Stress at Elevated Temperatures, *J. Iron Steel Inst.*, v. 122, 1930, pp. 215–247.
377. HESSENMÜLLER, J. H.: Bestimmung der Oberflächenhärte und der Einsatztiefe an gehärteten Werkstücken mit dem Härteprüfer "Testor" (Determination of Surface Hardness and of Depth of Case of Hardened Work Pieces with Hardness Tester "Testor"), *Werkzeugmaschine*, v. 34, No. 3, 1930, pp. 41–45. Abst.: *Automotive Abst.*, v. 7, 1930, p. 75.
378. HURST, J. E.: Chemical Composition and Mechanical Properties of Centrifugal Castings, *Metallurgia*, v. 2, 1930, pp. 13–16, 53–56.
379. JACOBUS, D. S.: Working Stresses for Steel at High Temperatures, *Trans. Am. Soc. Mech. Eng.*, v. 52, 1930, FSP-52-35, pp. 295–299.
380. KANTER, J. J., and L. W. SPRING: Some Long-time Tension Tests of Steels at Elevated Temperatures, *Proc. Am. Soc. Test. Mat.*, v. 30, part 1, 1930, pp. 110–132.
381. KEIL, O. VON: Die Graphitbildung im Gusseisen (Graphite Formation in Cast Iron), *Archiv f. d. Eisenhüttenwesen*, v. 4, 1930, pp. 245–250.
382. KLINGENSTEIN, T.: Einfluss des Gefügezustandes und der Zusammensetzung von Gusseisen auf die Verschleissfestigkeit, mit besonderer Berücksichtigung des Phosphorgehaltes (Influence of Structure and Composition, Especially Phosphorus Content, on Wear Resistance of Cast Iron), *Mitt. Forsch.-Anst. G.H.H. Konzerns*, v. 1, 1930, pp. 18–24. Abst.: *Metals & Alloys*, v. 2, 1931, p. 171.
383. KNACKSTEDT, W.: "Die Werkstoffdämpfung bei Drehschwingungen nach dem Dauerprüfverfahren und dem Ausschwingverfahren" (Vibration Damping of Materials in Endurance and Vibration Testing), NEM Verlag, Berlin, 1930, 60 pp.
384. KOMMERS, J. B.: The Effect of Understressing on Cast Iron and Open-hearth Iron, *Proc. Am. Soc. Test. Mat.*, v. 30, part 2, 1930, pp. 368–383.
385. KÖTTGEN, E.: Über Dauerformen und den Einfluss der Abkühlgeschwindigkeit bei Grauguss (Permanent Molds and the Influence of Cooling Rate on Gray Cast Iron), *Giesserei*, v. 17, 1930, pp. 1061–1064, 1089–1095.
386. KRYNITSKY, A. I., and W. N. HARRISON: Blistering Phenomena in the Enameling of Cast Iron, *Bur. Stand. J. Research*, v. 4, 1930, pp. 757–807. Abst.: *Metals & Alloys*, v. 1, 1930, pp. 663–666. Also: *J. Am. Ceramic Soc.*, v. 13, 1930, pp. 16–61.
387. LEE, P. F., and H. A. SCHADE: Structural Steel Heat Treated, *Iron Age*, v. 125, 1930, pp. 510–514.

388. McAdam, D. J., Jr.: Influence of Cyclic Stress on Corrosion, Am. Inst. Min. Met. Eng., *Tech. Pub.* 329, 1930, 42 pp.

389. McAdam, D. J., Jr.: The Influence of Stress Range and Cycle Frequency on Corrosion, *Proc. Am. Soc. Test. Mat.*, v. 30, part 2, 1930, pp. 411–447.

390. McAdam, D. J., Jr.: Stress and Corrosion, 3d Int. Cong. Applied Mechanics, Stockholm, 1930, *Proc.*, part 2, pp. 269–293.

391. McCrae, J. V., and R. L. Dowdell: Preliminary Studies of the Effect of Deoxidation and Mold Conditions on the Tensile Properties of Carbon Steel Castings, *Bur. Stand. J. Research*, v. 5, 1930, pp. 265–277.

392. McCrae, J. V., and R. L. Dowdell: The Effect of Deoxidation and Mold Conditions on the Tensile Properties of Carbon Steel Castings, *Trans. Am. Soc. Steel Treat.*, v. 18, 1930, pp. 159–211.

393. Mitchell, H. A.: Getting the Most out of Steel Castings, *Iron Age*, v. 126, 1930, pp. 914–917.

394. Moffett, R. W.: Effect of Low Temperatures upon the Impact Resistance of Steel Castings, *Can. J. Research*, v. 2, 1930, pp. 327–340.

395. Moisseiff, L. S.: Investigation of Cold-drawn Bridge Wire, *Proc. Am. Soc. Test. Mat.*, v. 30, part 2, 1930, pp. 313–349.

396. Moldenke, R.: "The Principles of Iron Founding," McGraw-Hill Book Company, Inc., New York, 2d ed., 1930, 654 pp.

397. Moore, H. F., and T. Ver: A Study of Slip Lines, Strain Lines, and Cracks in Metals under Repeated Stress, Univ. of Illinois, Exp. Sta., *Bull.* 208, 1930, 60 pp.

398. Norbury, A. L., and E. Morgan: The Effect of Melting Conditions on the Microstructure and Mechanical Strengths of Gray Cast Irons Containing Various Amounts of Carbon and Silicon, *J. Iron Steel Inst.*, v. 121, 1930, pp. 367–392.

399. O'Neill, H.: A New Method of Making Ball Hardness Tests of Metals; and a Plotting Table for the Simplification of "Meyer (Hardness) Analysis," *Iron Steel Inst., Carnegie Schol. Mem.*, v. 19, 1930, pp. 19–38.

400. Peterson, R. E.: Fatigue Tests of Small Specimens with Particular Reference to Size Effect, *Trans. Am. Soc. Steel Treat.*, v. 18, 1930, pp. 1041–1056.

401. Petrenko, S. N.: Relationships between Rockwell and Brinell Numbers, *Bur. Stand. J. Research*, v. 5, 1930, pp. 19–50. Extended Abst.: *Metals & Alloys*, v. 1, 1930, pp. 782–784.

402. Pomp, A., and W. Enders: Zur Bestimmung der Dauerstandsfestigkeit im Abkürzungsverfahren (Determination of Creep by an Accelerated Method), *Mitt. K.-W. Inst. Eisenforschung*, v. 12, 1930, pp. 127–147.

403. Pomp, A., and A. Lindeberg: Festigkeitseigenschaften und Gefügeausbildung von gezogenem Stahldraht in Abhängigkeit von der voraufgegangenen Wärmebehandlung (The Effect of Variations in the Patenting Treatment on the Toughness and Structure of Cold-drawn

Wire), *Mitt. K.-W. Inst. Eisenforschung*, v. 12, 1930, pp. 39–54. Also: *Stahl u. Eisen*, v. 50, 1930, pp. 1462–1467.

404. Pullin, V. E.: X-rays in Engineering Practice, *Proc. Inst. Mech. Eng.*, 1930, part 2, pp. 1133–1157. Abst.: *Engineer*, v. 150, 1930, pp. 686–688.

405. Rohn, W.: Creep Stress Determination, *Metallurgist*, v. 6, 1930, pp. 22–23.

406. Roš, M.: Die breitflanschigen Differdinger-Grey-Träger (Heavy I-Beams), Eidgenössische Materialprüfungsanstalt an der E.T.H. Zürich, *Rep.* 52, 1930, 34 pp.

407. Saeger, C. M., Jr., and E. J. Ash: Methods for Determining the Volume Changes Undergone by Metals and Alloys during Casting, *Trans. Am. Foundrymen's Assoc.*, v. 38, 1930, pp. 107–145.

408. Sauveur, A.: Steel at Elevated Temperatures, *Trans. Am. Soc. Steel Treat.*, v. 17, 1930, pp. 410–448.

409. Schoenmaker, P.: Zerreiss- und Schlag-Versuche bei tiefen Temperaturen (Tensile and Impact Tests at Low Temperatures), First Communication, New Int. Assoc. Test. Mat., Group A, 1930, pp. 237–250.

410. Schüz, E., and R. Stotz: "Der Temperguss" (Malleable Iron), Julius Springer, Berlin, 1930, 390 pp.

411. Schwartz, H. A.: Factors Affecting the Machinability of Malleable Cast Iron, *Trans. Am. Foundrymen's Assoc.*, v. 38, 1930, pp. 210–258.

412. Schwarz, M. von, and A. Väth: Schleuderguss und seine metallkundliche Untersuchung (Centrifugal Castings and Their Metallurgical Investigation), *Giesserei*, v. 17, 1930, pp. 177–182, 204–208, 230–234, 253–259.

413. Siebel, E., and A. Pomp: Die Prüfung von Feinblechen durch den Tiefzieh-Weitungsversuch (Testing Thin Sheets by the Deep-drawing Test), *Mitt. K.-W. Inst. Eisenforschung*, v. 12, 1930, pp. 115–125.

414. Sjovall, H.: Magnetic Hardness Test for Ball Races, *Metallurgist*, v. 6, 1930, pp. 121–122.

415. Spring, L. W.: Some Considerations and Tests for Cast Materials for High-temperature High-pressure Service, *Proc. Inst. Br. Foundrymen*, v. 24, 1930–1931, pp. 239–289, 305–311. Also: *Foundry Tr. J.*, v. 44, 1931, pp. 399–404; v. 45, 1931, pp. 7–10, 12. Also: *Trans. and Bull. Am. Foundrymen's Assoc.*, v. 2, No. 10, 1931, pp. 13–55.

416. Tamaru, K., and S. Sekito: On the Quantitative Determination of Retained Austenite in Quenched Steels, *Sci. Rep.*, Sendai, ser. 1, v. 20, 1930, pp. 377–394.

417. Thum, A., and H. Holdt: Die Dauerstandfestigkeit von Stahlguss bei erhöhten Temperaturen (Creep of Cast Steel at Elevated Temperatures), *Giesserei*, v. 17, 1930, pp. 333–339.

418. Vanick, J. S., and P. D. Merica: Corrosion- and Heat-resistant Nickel-copper-chromium Cast Iron, *Trans. Am. Soc. Steel Treat.*, v. 18, 1930, pp. 923–942.

419. VOLLMER, L. W., and B. B. WESTCOTT: Effect of Hydrogen Sulphide on Wire Rope, *Trans. Am. Soc. Mech. Eng.*, v. 52, 1930, PET-52-7, pp. 39–45.

420. WALLICHS, A., and H. DABRINGHAUS: Die Zerspanbarkeit des Gusseisens im Drehvorgang (Machinability of Cast Iron in Lathe Tests), *Giesserei*, v. 17, 1930, pp. 1169–1177, 1197–1201.

421. WALLICHS, A., and H. DABRINGHAUS: Die Zerspanbarkeit und die Festigkeitseigenschaften bei Stahl und Stahlguss (Machinability and Strength of Wrought and Cast Steels), *Maschinenbau*, v. 9, 1930, pp. 257–262.

422. WALLICHS, A., and K. KREKELER: Die Prüfung der Zerspanbarkeit mittels des Leyensetter-Pendels (Testing Machinability with the Leyensetter Pendulum), *Archiv f. d. Eisenhüttenwesen*, v. 4, 1930, pp. 41–43.

423. WALLICHS, A., and K. KREKELER: Stahlguss und Zerspanbarkeit (Cast Steel and Machinability), *Giesserei*, v. 17, 1930, pp. 325–329.

424. WALLICHS, A., and H. OPITZ: Die Prüfung der Zerspanbarkeit von Automatenstahl (Testing the Machinability of Free-cutting Steel), *Archiv f. d. Eisenhüttenwesen*, v. 4, 1930, pp. 251–260.

425. WALLS, F. J., and A. HARTWELL, JR.: Some Phases of Heat Treatment of Cylinder and Alloy Irons, *Trans. Am. Foundrymen's Assoc.*, v. 38, 1930, pp. 865–896.

426. ZIMMERLI, F. P., W. P. WOOD, and G. D. WILSON: The Effect of Temperature upon the Torsional Modulus of Spring Materials, *Proc. Am. Soc. Test. Mat.*, v. 30, part 2, 1930, pp. 350–367.

1931

427. ABEGG, R., F. AUERBACH, and I. KOPPEL (editors): "Handbuch der anorganischen Chemie" (Handbook of Inorganic Chemistry), S. Hirzel, Leipzig, v. 4, section 3, part 2A, 1931, p. 238.

428. ALLEN, R. J.: Discussion of paper by J. W. Bolton and H. Bornstein on Effect of Elevated Temperatures on Certain Mechanical Properties of Gray Cast Iron and Malleable Iron, "Symposium on Effect of Temperature on Metals," Am. Soc. Test. Mat. and Am. Soc. Mech. Eng., Philadelphia and New York, 1931, pp. 459–463.

429. American Foundrymen's Association and American Society for Testing Materials: Symposium on Malleable Iron Castings, *Proc. Am. Soc. Test. Mat.*, v. 31, part 2, 1931, pp. 317–434.

430. American Society of Mechanical Engineers: Supplement to a Bibliography on Effect of Temperature on Properties of Metals together with a Bibliography on Oxidation of Metals at High Temperatures, A.S.M.E. *Research Publ.*, 1931, 27 pp.

431. American Society of Mechanical Engineers and American Society for Testing Materials: "Symposium on Effect of Temperature on the Properties of Metals," The Societies, New York and Philadelphia, 1931, 829 pp.

432. AUSTIN, C. R.: Effect of Surface Decarburization on Fatigue Properties of Steel, *Metals & Alloys*, v. 2, 1931, pp. 117–119.

433. Bailey, R. W., J. H. S. Dickenson, N. P. Inglis, and J. L. Pearson: The Trend of Progress in Great Britain on the Engineering Use of Metals at Elevated Temperatures, American Society of Mechanical Engineers and American Society for Testing Materials, "Symposium on Effect of Temperature on the Properties of Metals," 1931, pp. 218–244.

434. Bardenheuer, P., and W. Bottenberg: Einfluss der Schmelzbehandlung auf den Gasgehalt und die Schwindung von weissem und grauem Gusseisen (Influence of Melting Practice on the Gas Content and Shrinkage of White and Gray Cast Irons), *Mitt. K.-W. Inst. Eisenforschung*, v. 13, 1931, pp. 149–159. Also: *Giesserei*, v. 19, 1932, pp. 201–206.

435. Becket, F. M.: Manganese-free Zirconium-treated Steels, *Mining and Metallurgy*, v. 12, 1931, pp. 234–236.

436. Bleakney, H. H.: Machining of Steel Is Conditioned by Quality of Metal, *Iron Age*, v. 128, 1931, pp. 1608–1609, 1662.

437. Bolton, J. W., and H. Bornstein: Effect of Elevated Temperatures on Certain Mechanical Properties of Gray Cast Iron and Malleable Iron, American Society of Mechanical Engineers and American Society for Testing Materials, "Symposium on Effect of Temperature on the Properties of Metals," 1931, pp. 436–465.

438. Bornstein, H.: A Comparison of Various Sizes of Test Bars Representing Cast Irons from Five Foundries, *Proc. Am. Soc. Test. Mat.*, v. 31, part 1, 1931, pp. 152–159.

439. Boston, O. W.: An Investigation of Methods to Determine the Machinability of Malleable Iron Castings, *Proc. Am. Soc. Test. Mat.*, v. 31, part 2, 1931, pp. 388–421.

440. Boston, O. W.: Machining Properties of Some Cold-drawn Steel, *Trans. Am. Soc. Mech. Eng.*, v. 53, 1931, MSP 53–6, pp. 41–55.

441. Briggs, C. W., and R. A. Gezelius: Detecting Defects by Radiography, Using Gamma Rays, *Trans. Am. Foundrymen's Assoc.*, v. 39, 1931, pp. 65–92.

442. Bull, R. A.: Chemical Composition as Employed for Classifying Carbon and Alloy Steel for Castings, *Trans. Am. Foundrymen's Assoc.*, v. 39, 1931, pp. 257–274. Abst.: *Iron Age*, v. 128, 1931, pp. 174–175, 207.

443. Bull, R. A.: The Performances of Cast Carbon and Low-alloy Steels in High- and Low-temperature Service, American Society of Mechanical Engineers and American Society for Testing Materials, "Symposium on Effect of Temperature on the Properties of Metals," 1931, pp. 394–435.

444. Bull, R. A.: Steel Castings as Machine Parts, *Machine Design*, v. 3, 1931, No. 4, pp. 36–39; No. 5, pp. 37–40.

445. Coyle, F. B.: Heat Treatment Fundamentals of Plain and Alloy Cast Iron, *Metals & Alloys*, v. 2, 1931, pp. 120–131.

446. Diepschlag, E., and F. Eggert: Die Schneidfähigkeit von Grauguss in Abhängigkeit von Silizium- und Kohlenstoffgehalt (Effect of Silicon and Carbon on the Machinability of Gray Iron), *Giesserei*, v. 18, 1931, pp. 313–318.

447. DIGGES, T. G.: Influence of Chemical Composition and Heat Treatment of Steel Forgings on Machinability with Shallow Lathe Cuts, *Bur. Stand. J. Research*, v. 6, 1931, pp. 977–992. Also: *Trans. Am. Soc. Mech. Eng.*, v. 54, 1932, MSP 54–4, pp. 49–59.

448. DIXON, E. S.: Needs of the Oil Industry for Metals at High Temperatures, American Society of Mechanical Engineers and American Society for Testing Materials, "Symposium on Effect of Temperature on the Properties of Metals," 1931, pp. 66–80.

449. FENSTERMACHER, J. D.: Improvements in Physical Properties in Large Carbon and Alloy Castings, *Steel Founder*, v. 2, June–July, 1931, pp. 4–7.

450. FRANCE, R. D.: Endurance Testing of Steel: Comparison of Results Obtained with Rotating-beam versus Axially-loaded Specimens, *Proc. Am. Soc. Test. Mat.*, v. 31, part 2, 1931, pp. 176–193.

451. Franklin Institute: Committee on Science and the Arts: Pendulum Hardness Tester, *J. Franklin Inst.*, v. 212, 1931, pp. 376–386.

452. FRENCH, H. J., W. KAHLBAUM, and A. A. PETERSON: Flow Characteristics of Special Fe-Ni-Cr Alloys and Some Steels at Elevated Temperatures, *Trans. Am. Soc. Mech. Eng.*, v. 53, 1931, FSP-53-9, pp. 97–130.

453. GILLETT, H. W.: Testing Deep Drawing Qualities of Sheet Metal, *Metals & Alloys*, v. 2, 1931, pp. 214–222; discussion by C. L. Eksergian, pp. 220–222.

454. HANKINS, G. A., and M. L. BECKER: The Effect of Surface Conditions Produced by Heat Treatment on the Fatigue Resistance of Spring Steels, *J. Iron Steel Inst.*, v. 124, 1931, pp. 387–460.

455. HARDER, O. E.: Nitriding for the Engineer, *Metals & Alloys*, v. 2, 1931, pp. 132–142.

456. HEYDEKAMPF, G. S. VON: Damping Capacity of Materials, *Proc. Am. Soc. Test. Mat.*, v. 31, part 2, 1931, pp. 157–171.

457. HURST, J. E.: Centrifugal Casting of Metals and Alloys, *Metals & Alloys*, v. 2, 1931, pp. 197–205.

458. INGLIS, N. P., and G. F. LAKE: Corrosion-fatigue Tests of Mild Steel and Chromium-nickel Austenitic Steel in River Tees Water, *Trans. Faraday Soc.*, v. 27, 1931, pp. 803–808.

459. Iron and Steel Institute: "The First Report of the Corrosion Committee," The Institute, London, 1931, 250 pp.

460. ISEMER, H.: Die Steigerung der Schwingungsfestigkeit von Gewinden durch Oberflächendrücken (Increasing the Endurance Limit of Threads by Compression of the Surface), *Mitt. d. Wöhler-Inst., Braunschweig*, No. 8, NEM Verlag, Berlin,1931, 64 pp.

461. JOMINY, W. E.: The Surface Decarburization of Steel at Heat Treating Temperatures, Univ. of Michigan, Dept. of Eng. Research, *Bull.* 18, 1931, 49 pp.

462. JORDAN, L.: The Wear of Metals, *Mech. Eng.*, v. 53, 1931, pp. 644–650.

463. JÜNGER, A.: Weitere Erfahrungen in der Prüfung der Dauerfestigkeit verschiedener Werkstoffe auf der M.A.N. Biegeschwingungsmaschine

(Further Work on Endurance Testing), *Mitt. Forsch.-Anst. G.H.H. Konzerns*, v. 1, 1931, pp. 45–58.

464. KALLEN, H., and H. SCHRADER: Die Durchvergütung von Konstruktionsstählen unter Berücksichtigung des Einflusses von Stückquerschnitt und Legierung (Deep Hardening of Structural Steels as Affected by Cross-section and Composition), *Archiv f. d. Eisenhüttenwesen*, v. 4, 1931, pp. 383–392. Abst.: *Stahl u. Eisen*, v. 51, 1931, p. 387.

465. KAUFMANN, E.: "Über die Dauerbiegefestigkeit einiger Eisenwerkstoffe und ihre Beeinflussung durch Temperatur und Kerbwirkung" (Effect of Temperature and Notches on the Endurance Limit of Some Ferrous Materials), Julius Springer, Berlin, 1931, 89 pp.

466. KLINGENSTEIN, T.: Weitere Untersuchungen über den Einfluss der Zusammensetzung und des Gefügeaufbaus auf den Verschleiss von Gusseisen (Effect of Composition and Structure on the Wear of Cast Iron), *Mitt. Forsch.-Anst. G.H.H.Konzerns*, v. 1, 1931, pp. 80–92. Abst.: *Metals & Alloys*, v. 2, 1931, p. 154.

467. LAURSON, P. G., and W. J. COX: "Properties and Mechanics of Materials," John Wiley & Sons, Inc., New York, 1931, 353 pp.

468. LOGAN, K. H., and V. A. GRODSKY: Soil Corrosion Studies, 1930—Rates of Corrosion and Pitting of Bare Ferrous Specimens, *Bur. Stand. J. Research*, v. 7, 1931, pp. 1–35.

469. LUCAS, F. F.: On the Art of Metallography, *Trans. Am. Inst. Min. Met. Eng.*, v. 95, 1931, pp. 11–44.

470. MACKENZIE, J. T.: Tests on Cast Iron Specimens of Various Diameters, *Proc. Am. Soc. Test. Mat.*, v. 31, part 1, 1931, p. 160.

471. MACQUIGG, C. E.: High-chromium Steels for Extreme Service Conditions, American Society of Mechanical Engineers and American Society for Testing Materials, "Symposium on Effect of Temperature on the Properties of Metals," 1931, pp. 589–609.

472. MCADAM, D. J., JR.: Stress Corrosion of Metals, Int. Assoc. Test. Mat., Cong. Zürich, 1931, v. 1, pp. 228–246.

473. MCADAM, D. J., JR.: Influence of Water Composition on Stress Corrosion, *Proc. Am. Soc. Test. Mat.*, v. 31, part 2, 1931, pp. 259–278.

474. MOCHEL, N. L.: Thermal Expansion of Metals, American Society of Mechanical Engineers and American Society for Testing Materials, "Symposium on Effect of Temperature on the Properties of Metals," 1931, pp. 683–724.

475. MONEYPENNY, J. H. G.: "Stainless Iron and Steels," John Wiley & Sons, Inc., New York, 2d ed., 1931, 304 pp.

476. MOORE, H. F.: Crackless Plasticity—A New Property of Metals, *Iron Age*, v. 128, 1931, pp. 674–677, 721.

477. MURPHY, D. W., W. P. WOOD, and W. E. JOMINY: Scaling of Steel at Elevated Temperatures by Reaction with Gases and the Properties of the Resulting Oxides, *Trans. Am. Soc. Steel Treat.*, v. 19, 1931, pp. 193–232.

478. PALMAER, K. W.: "Corrosion of Metals," Ingeniörsvetenskapsakademiens, Stockholm, No. 93, 1929, 347 pp., No. 108, 1931, 198 pp.

479. PFEIL, L. B.: The Constitution of Scale, *J. Iron Steel Inst.*, v. 123, 1931, pp. 237–258.

480. PULLIN, V. E.: X-ray Examination of Castings, *Foundry Tr. J.*, v. 44, 1931, pp. 23–24.

481. RABALD, E.: "Werkstoffe und Korrosion" (Materials and Corrosion), Otto Spamer, Leipzig, 1931, 2 vols.

482. REMMERS, W. E.: The Permanent Growth of Gray Cast Iron, *Trans. Am. Inst. Min. Met. Eng.*, v. 95, 1931, pp. 219–229.

483. ROSENBERG, S. J.: How Carbon Content and Heat Treatment Can Affect Wear Resistance, *Iron Age*, v. 128, 1931, pp. 1366–1367.

484. RUSSELL, H. W.: Effect of Low Temperatures on Metals and Alloys, American Society of Mechanical Engineers and American Society for Testing Materials, "Symposium on Effect of Temperature on the Properties of Metals," 1931, pp. 658–682.

485. SACHS, G.: "Spanlose Formung der Metalle" (Chipless Forming of Metals), Julius Springer, Berlin, 1931, 127 pp.

486. SAEGER, C. M., JR., and A. I. KRYNITSKY: A Practical Method for Studying the Running Quality of a Metal Cast in Foundry Molds, *Trans. Am. Foundrymen's Assoc.*, v. 39, 1931, pp. 513–540.

487. SALMON, E. H.: "Materials and Structures," Longmans, Green & Company, London, 1931, v. 1, 638 pp.

488. SAUERWALD, F., B. SCHMIDT, and G. KRÄMER: Über den Sprödigkeitsbereich von Eisen bei tiefen Temperaturen (The Range of Brittleness of Iron at Low Temperatures), *Z. f. Physik*, v. 67, 1931, pp. 179–183.

489. SCHWARTZ, H. A.: Malleable Iron, Recent Progress, *Metals & Alloys*, v. 2, 1931, pp. 143–149.

490. SCOTT, H.: High-temperature Characteristics of Metals Revealed by Bending, *Proc. Am. Soc. Test. Mat.*, v. 31, part 2, 1931, pp. 129–156.

491. SHELTON, S. M.: Fatigue Testing of Wire, *Proc. Am. Soc. Test. Mat.*, v. 31, part 2, 1931, pp. 204–220.

492. SPOONER, A. P., and F. B. FOLEY: High-temperature Creep Properties of Wrought Carbon and Low-alloy Steels, American Society of Mechanical Engineers and American Society for Testing Materials, "Symposium on Effect of Temperature on the Properties of Metals," 1931, pp 368–393.

493. STEINBERG, S., and W. SUBOW: Über die Alterung des gehärteten Kohlenstoffstahles (Aging of Hardened Carbon Steel), *Stahl u. Eisen*, v. 51, 1931, pp. 911–913.

494. TAMARU, K., and S. SEKITO: On the Quantitative Determination of Retained Austenite in Quenched Steels, *Sci. Rep.*, Sendai, ser. 1, v. 20, 1931, pp. 377–394.

495. TANIMURA, H.: Influences of Cooling Velocity and Melting Temperature on the Graphitization of Cast Iron, *Mem. Coll. Eng., Kyushu Imp. Univ.*, v. 6, No. 2, 1931, pp. 115–188.

496. TAPSELL, H. J.: "Creep of Metals," Oxford University Press, London, 1931, 285 pp.

497. VAN DUSEN, M. S.: Note on Applications of Data on the Thermal Conductivity of Metals. American Society of Mechanical Engineers

and American Society for Testing Materials, "Symposium on Effect of Temperature on the Properties of Metals," 1931, pp. 725–729.

498. VAN WERT, L. R.: Some Notes on Blue Brittleness, *Trans. Am. Inst. Min. Met. Eng.*, v. 95, 1931, pp. 230–246.

499. WHITNEY, H. L.: Discussion of paper by A. P. Spooner and F. B. Foley on High-temperature Creep Properties of Wrought Carbon and Low-alloy Steels, American Society of Mechanical Engineers and American Society for Testing Materials, "Symposium on Effect of Temperature on the Properties of Metals," 1931, p. 391.

500. WOLF, F. L., and L. A. MEISSE: Corrosion of Malleable Iron, *Proc. Am. Soc. Test. Mat.*, v. 31, part 2, 1931, pp. 422–434.

501. YENSEN, T. D., and N. A. ZIEGLER: Magnetic Properties versus Allotropic Transformations of Iron Alloys, *Trans. Am. Inst. Min. Met. Eng.*, v. 95, 1931, pp. 313–324.

502. ZEYEN, K. L.: Über Festigkeitswerte von stehend und liegend gegossenen Gusseisen-Probestäben (The Strength of Cast-iron Test Pieces Cast Vertically and Horizontally), *Giesserei*, v. 18, 1931, pp.177–180.

1932

503. American Foundrymen's Association: Reports of Gray Iron Committees on Recommended Practice. Analyses for Classes of Castings, *Trans. Am. Foundrymen's Assoc.*, v. 39, 1932, pp. 59–64.

504. American Society for Testing Materials: Summary of Present-day Knowledge of Fatigue Phenomena in Metals, *Proc. Am. Soc. Test. Mat.*, v. 32, part 1, 1932, pp. 139–147.

505. American Society of Mechanical Engineers and American Society for Testing Materials: Report of Joint Research Committee on Effect of Temperature on the Properties of Metals, *Proc. Am. Soc. Test. Mat.*, v. 32, part 1, 1932, pp. 148–192.

506. ASH, E. J., and C. M. SAEGER, JR.: Shop Method for Determining Volume Changes in Cast Iron during Casting, *Trans. Am. Foundrymen's Assoc.*, v. 40, 1932, pp. 188–200.

507. BAILEY, R. W., and A. M. ROBERTS: Testing of Materials for Service in High-temperature Steam Plant, *Engineering*, v. 133, 1932, pp. 261–265, 295–298.

508. BAIN, E. C.: Factors Affecting the Inherent Hardenability of Steel, *Trans. Am. Soc. Steel Treat.*, v. 20, 1932, pp. 385–428.

509. BARR, W., and W. E. BARDGETT: An Accelerated Test for the Determination of the Limiting Creep Stress of Metals, *Proc. Inst. Mech. Eng.*, v. 122, 1932, pp. 285–373.

510. BOLTON, J. W.: Some Notes on Mechanism of Deformations in Gray Iron, *Proc. Am. Soc. Test. Mat.*, v. 32, part 2, 1932, pp. 477–491.

511. BROPHY, G. R.: Damping Capacity of Steels and Correlation with Other Properties, *Iron Age*, v. 130, 1932, pp. 800–802, 989; Adv. p. 10.

512. BULL, R. A.: Purchase Requirements for Steel Castings, with Notes on Physical Properties in Test Bars and in Commercial Castings, *Proc. Am. Soc. Test. Mat.*, v. 32, part 2, 1932, pp. 77–113. Also in:

American Foundrymen's Association and American Society for Testing Materials, "Symposium on Steel Castings," 1932, pp. 39–75.

513. Cioffi, P. P.: Hydrogenized Iron, *Physical Rev.*, ser. 2, v. 39, 1932, pp. 363–367.

514. Coffin, F. P., and T. H. Swisher: Flow of Steels at Elevated Temperatures, *Trans. Am. Soc. Mech. Eng.*, v. 54, 1932, APM-54-6, pp. 59–68.

515. Cunningham, J. F., and J. Gilchrist: Impact Characteristics of Steel Rails at Low Temperatures, *Trans. Am. Soc. Steel Treat.*, v. 19, 1932, pp. 624–638.

516. Digges, T. G.: Effect of Lathe Cutting Conditions on the Hardness of Carbon and Alloy Steels, Am. Soc. Mech. Eng., *Preprint* 13, section MSP, 1932, 12 pp. Abst.: *Bur. Stand. J. Research*, v. 10, 1933, pp. 77–78.

517. Donaldson, J. W.: The Thermal Conductivities of Ingot Mold Irons, Iron Steel Inst., Fourth Report on the Heterogeneity of Steel Ingots, *Special Rep.* 2, 1932, pp. 151–161.

518. Dwyer, P.: Makes Permanent Mold Castings, *Foundry*, v. 60, 1932, pp. 18–21, 72.

519. Ellis, O. W.: Forgeability of Steel as Influenced by Composition and Manufacture, *Metal Progress*, v. 22, No. 3, 1932, pp. 19–23. Also: *Trans. Am. Soc. Steel Treat.*, v. 21, 1933, pp. 673–707.

520. Epstein, S.: Embrittlement of Hot-galvanized Structural Steel, *Proc. Am. Soc. Test. Mat.*, v. 32, part 2, 1932, pp. 293–379.

521. Garre, B., and E. Grathoff: Einfluss des Glühens auf die Drehschwingungsfestigkeit von Stahlguss (Influence of Annealing on the Torsional Endurance Strength of Steel), *Stahl u. Eisen*, v. 52, 1932, p. 493.

522. Gillett, H. W.: The Resistance of Copper and Its Alloys to Repeated Stress, *Metals & Alloys*, v. 3, 1932, pp. 200–204, 236–238, MA314; pp. 257–262, MA325; pp. 275–280.

523. Gough, H. J.: Corrosion Fatigue of Metals, *J. Inst. Metals*, v. 49, 1932, pp. 17–92.

524. Gough, H. J., and D. G. Sopwith: Atmospheric Action as a Factor in Fatigue of Metals, *J. Inst. Metals*, v. 49, 1932, pp. 93–122.

525. Gough, H. J., and D. G. Sopwith: The Behaviour of a Single Crystal of Aluminium under Alternating Torsional Stresses while Immersed in a Slow Stream of Tap Water, *Proc. Royal Soc.*, London, v. 135A, 1932, pp. 392–411.

526. Hamilton, W. C.: General Survey of the Steel Castings Industry, American Foundrymen's Association and American Society for Testing Materials, "Symposium on Steel Castings," The Societies, Chicago and Philadelphia, 1932, pp. 7–11.

527. Hankins, G. A., and M. L. Becker: The Fatigue Resistance of Unmachined Forged Steels, *J. Iron Steel Inst.*, v. 126, 1932, pp. 205–236.

528. Harvey, W. E.: Cadmium Plating vs. Corrosion Fatigue—Pickling vs. Corrosion Fatigue, *Metals & Alloys*, v. 3, 1932, pp. 69–72.

529. HEDGES, E. S.: "Protective Films on Metals," Chapman and Hall, Ltd., London, 1932, 276 pp.

530. HEINRICH, F. H.: Einfluss der Lage des Probestabes im Stahlgussstücke auf die Zerreisswerte (Influence of the Position of the Test Piece in a Steel Casting on the Tensile Strength), *Stahl u. Eisen*, v. 52, 1932, pp. 1017–1020.

531. HELLER, P. A.: Die Wärmebehandlung des grauen Gusseisens (Heat Treatment of Gray Cast Iron), *Giesserei*, v. 19, 1932, pp. 413–419.

532. HIGHRITER, H. W.: Short-cycle Malleablizing, *Trans. Am. Foundrymen's Assoc.*, v. 40, 1932, pp. 72–87.

533. HONEGGER, E.: Der Elastizitätsmodul von Stahl bei hohen Temperaturen (Modulus of Elasticity of Steel at High Temperatures), *B B C Nachrichten*, v. 19, 1932, pp. 61–64. Also: *Mech. World and Eng. Record*, v. 92, 1932, pp. 357–358.

534. INGLIS, N. P., and G. F. LAKE: Corrosion Fatigue Tests of Nitrided Steel and Nickel-plated Steel in River Tees Water, *Trans. Faraday Soc.*, v. 28, 1932, pp. 715–721.

535. ITIHARA, M.: Impulsive Torsion Tests, *J. Soc. Mech. Eng.*, Tokyo, v. 35, 1932, pp. 678–691.

536. JUDSON, H. H.: High-test Iron for Pressure Castings, *Trans. Am. Foundrymen's Assoc.*, v. 40, 1932, pp. 153–167.

537. KIKUTA, T.: Effect of Thermal Treatment on the Characteristics of Black-heart Malleable Iron, *Trans. Am. Foundrymen's Assoc.*, v. 40, 1932, pp. 401–443.

538. KOMMERS, J. B.: Understressing and Notch Sensitiveness in Fatigue, *Eng. News-Record*, v. 109, 1932, pp. 353–355.

539. KÖSTER, W., H. VON KÖCKRITZ, and E. H. SCHULZ: Zur Kenntnis der Form der Spannungs-Dehnungs-Kurven auf Grund der Messung des zeitlichen Verlaufes der Alterung weichen Stahles (Form of the Tension-elongation Curves on the Basis of Measurement of the Temporary Course of Aging of Soft Steel), *Archiv f. d. Eisenhüttenwesen*, v. 6, 1932, pp. 55–60. Abst.: *Metals & Alloys*, v. 4, 1933, p. MA179.

540. KUNTZE, W.: Die Praxis des Kerbzugversuchs im Hinblick auf die technische Kohäsionsermittlung (The Practice of the Notched Tensile Test with Reference to the Technical Determination of Cohesion in Metals), *Metallwirtschaft*, v. 11, 1932, pp. 179–184.

541. KUNTZE, W.: Die Methodik der technischen Kohäsionsermittlung (The Method for the Technical Determination of Cohesion), *Metallwirtschaft*, v. 11, 1932, pp. 343–348.

542. LIGHTNER, M. W., and C. H. HERTY, JR.: The Transverse Impact Strength of Plain Carbon, Normalized Steels, *Cooperative Bull.* 59, Min. Met. Advisory Boards to Carnegie Inst. Tech. and Bur. Mines, Pittsburgh, 1932, 33 pp.

543. LORENZ, A. W.: Problems and Practices in the Heat Treatment of Steel Castings, *Proc. Am. Soc. Test. Mat.*, v. 32, part 2, 1932, pp. 253–268. Also in: American Foundrymen's Association and American Society for Testing Materials, "Symposium on Steel Castings," 1932, pp. 215–230.

544. Lorenz, F. A., Jr.: Notes on the Design of Steel Castings, *Proc. Am. Soc. Test. Mat.*, v. 32, part 2, 1932, pp. 58–76. Also in: American Foundrymen's Association and American Society for Testing Materials, "Symposium on Steel Castings," 1932, pp. 20–38.

545. Lorig, C. H., and C. E. Williams: Physical and Mechanical Properties of Some Well-known Cast Steels, American Foundrymen's Association and American Society for Testing Materials, "Symposium on Steel Castings," The Societies, Chicago and Philadelphia, 1932, pp. 76–158.

546. MacPherran, R. S.: High-test Cast Iron in the United States of America, *Foundry Tr. J.*, v. 47, 1932, pp. 16–19, 25.

547. Main, F. L.: Brake-drum Metallurgy, *S.A.E. J.*, v. 31, 1932, pp. 330–338.

548. McAdam, D. J., Jr.: Influence of Stress on Corrosion, *Trans. Am. Inst. Min. Met. Eng.*, v. 99, 1932, pp. 282–322.

549. Mehl, R. F.: Non-destructive Testing Methods in the Iron and Steel Industry, *Yearbook Am. Iron Steel Inst.*, v. 22, 1932, pp. 126–184.

550. Meller, K.: "Elektrische Lichtbogenschweissung" (Electric-arc Welding), S. Hirzel, Leipzig, 2d ed., 1932, 398 pp.

551. Mellor, J. W.: "A Comprehensive Treatise on Inorganic and Theoretical Chemistry," Longmans, Green & Company, London, 1932, v. 12, 944 pp.

552. Meyersberg, G.: Einfluss der Bearbeitung auf die Ergebnisse des Biegeversuchs bei Gusseisen (Influence of Machining on the Results of Bending Tests on Cast Iron), *Archiv f. d. Eisenhüttenwesen*, v. 5, 1932, pp. 511–512.

553. Middleton, T. R.: The High-frequency Furnace and Its Use for the Manufacture of Steel Castings, *Foundry Tr. J.*, v. 46, 1932, pp. 73–75.

554. Neuendorff, G. G.: Izett Steel, What It Is and What It Is Used For, *Metals & Alloys*, v. 3, 1932, pp. 61–68.

555. Pearce, J. G.: The Strength of Cast Iron in Relation to Combined Carbon Content, *Bull. Br. Cast Iron Research Assoc.*, v. 3, 1932, pp. 147–149. Abst.: *Metals & Alloys*, v. 4, 1933, p. MA310.

556. Peterson, R. E.: Fatigue of Shafts Having Keyways, *Proc. Am. Soc. Test. Mat.*, v. 32, part 2, 1932, pp. 413–420.

557. Pomp, A., and W. Höger: Dauerstandfestigkeitsuntersuchungen an Kohlenstoff- und niedriglegierten Stählen nach dem Abkürzungsverfahren (Creep Tests on Carbon and Low-alloy Steels by an Accelerated Method), *Mitt. K.-W. Inst. Eisenforschung*, v. 14, 1932, pp. 37–57.

558. Pullin, V. E.: Radium in Engineering Inspection, *Engineer*, v. 153, 1932, pp. 492–494.

559. Quick, G. W.: Tensile Properties of Rail Steels at Elevated Temperatures, *Bur. Stand. J. Research*, v. 8, 1932, pp. 173–189.

560. Quick, G. W.: The Resistance to Impact of Rail Steels at Elevated Temperatures, *Bur. Stand. J. Research*, v. 8, 1932, pp. 191–198.

561. Rosenhain, W.: The Testing of Castings, *J. Inst. Metals*, v. 48, 1932, pp. 273–310.

562. Ross, E. F.: Ford Improves Metal Fabricating Processes, *Steel*, v. 90, June 6, 1932, pp. 27–29.

563. Rötscher, R., and M. Fink: Beitrag zur Kerbzähigkeit hochwertiger Baustähle (Notch-impact Resistance of High-grade Structural Steels), *Z. Ver. deutsch. Ing.*, v. 76, 1932, pp. 173–177.

564. Saeger, C. M., Jr., and E. J. Ash: A Method for Determining the Volume Changes Occurring in Metals during Casting, *Bur. Stand. J. Research*, v. 8, 1932, pp. 37–60.

565. Saeger, C. M., Jr., and E. J. Ash: Volume Changes of Cast Irons during Casting, *Trans. Am. Foundrymen's Assoc.*, v. 40, 1932, pp. 172–187. Also: *Bur. Stand. J. Research*, v. 8, 1932, pp. 601–614.

566. Sandberg, C. P., O. F. A. Sandberg, and N. P. P. Sandberg: Effect of Controlled Cooling and Temperature Equalization on Internal Fissures in Rails, *Metals & Alloys*, v. 3, 1932, pp. 89–92.

567. Sawin, N. N., and E. Stachrowski: Comparative Tests of Hardened Steel by Means of Penetration Testing Apparatus, *Trans. Am. Soc. Steel Treat.*, v. 20, 1932, pp. 249–262.

568. Scheil, E.: Untersuchungen über das Wachsen von Gusseisen (The Growth of Cast Iron), *Archiv f. d. Eisenhüttenwesen*, v. 6, 1932, pp. 61–67.

569. Schroeder, W.: Untersuchungen über den Einfluss von Liegedauer, Temperatur, Gasgeschwindigkeit und Atmosphäre auf den Eisenabbrand (Influence of Time of Exposure, Temperature, Gas Velocity, and Atmosphere on Scaling), *Archiv f. d. Eisenhüttenwesen*, v. 6, 1932, pp. 47–54.

570. Sergeson, R.: Behavior of Some Irons and Steels under Impact at Low Temperatures, *Trans. Am. Soc. Steel Treat.*, v. 19, 1932, pp. 368–384.

571. Sims, C. E., and G. A. Lillieqvist: Inclusions—Their Effect, Solubility, and Control in Cast Steel, *Trans. Am. Inst. Min. Met. Eng.*, v. 100, 1932, pp. 154–195.

572. Swanger, W. H., and R. D. France: Effect of Zinc Coatings on the Endurance Properties of Steel, *Proc. Am. Soc. Test. Mat.*, v. 32, part 2, 1932, pp. 430–452.

573. Tammann, G.: "Lehrbuch der Metallkunde" (Textbook of Metallography), Leopold Voss, Leipzig, 4th ed., 1932, 536 pp.

574. Thum, A., and W. Buchmann: "Dauerfestigkeit und Konstruktion" (Endurance and Construction), VDI Verlag, Berlin, 1932, 81 pp.

575. Tour, S.: Heat Treating Furnace Atmospheres, *Metals & Alloys*, v. 3, 1932, pp. 270–272.

576. Voigt, E., and K. H. Christensen: Über die Dämpfungsfähigkeit und Schwingungsfestigkeit des Stahles (Damping Capacity and Fatigue Strength of Steel), *Mitt. K.-W. Inst. Eisenforschung*, v. 14, 1932, pp. 151–167.

577. Walle, R.: Festigkeitseigenschaften von Stahlguss bei tiefen Temperaturen (Tensile Characteristics of Cast Steels at Low Temperatures), *Stahl u. Eisen*, v. 52, 1932, pp. 489–490.

578. WALLICHS, A., and H. BEUTEL: Die Zerspanbarkeit des Stahlgusses im Bohrvorgang (The Machinability of Cast Steel in Drilling), *Giesserei*, v. 19, 1932, pp. 241–247.

579. WHITE, A. E., and R. SCHNEIDEWIND: The Metallurgy of Malleabilization, *Trans. Am. Foundrymen's Assoc.*, v. 40, 1932, pp. 88–124.

580. WIBERG, O. A.: High Pressures and High Temperatures for Steam Prime Movers, *Engineering*, v. 133, 1932, pp. 27–28.

581. YENSEN, T. D.: Effect of Impurities on Ferromagnetism, *Physical Rev.*, ser. 2, v. 39, 1932, pp. 358–363.

1933

582. American Foundrymen's Association and American Society for Testing Materials: Symposium on Cast Iron, *Proc. Am. Soc. Test. Mat.*, v. 33, part 2, 1933, pp. 115–273.

583. American Society of Mechanical Engineers and American Society for Testing Materials: Report of Joint Research Committee on Effect of Temperature on the Properties of Metals, *Proc. Am. Soc. Test. Mat.*, v. 33, part 1, 1933, pp. 209–251.

584. American Society for Steel Treating: Physical Property Charts for S.A.E. Steels, "National Metals Handbook," The Society, Cleveland, 1933, pp. 561–563.

585. American Society for Testing Materials: Report of Committee A-5 on Corrosion of Iron and Steel, *Proc. Am. Soc. Test. Mat.*, v. 33, part 1, 1933, pp. 140–172.

586. American Society for Testing Materials: Report of Subcommittee XV on Impact Testing, *Proc. Am. Soc. Test. Mat.*, v. 33, part 1, 1933, pp. 87–137.

587. American Society for Testing Materials: Standard Grain-size Chart for Classification of Steels, *A.S.T.M. Standards*, 1933, part 1, pp. 898–899.

588. American Society for Testing Materials: Standard Methods of Brinell Hardness Testing of Metallic Materials, *A.S.T.M. Standards*, 1933, part 1, pp. 942–948.

589. American Society for Testing Materials: Tentative Methods of Rockwell Hardness Testing of Metallic Materials, *Proc. Am. Soc. Test. Mat.*, v. 33, part 1, 1933, pp. 967–975.

590. AUSTIN, C. R., and J. R. GIER: Studies on a Modification of the Rohn Test for Investigating Creep of Metals, *Proc. Am. Soc. Test. Mat.*, v. 33, part 2, 1933, pp. 293–314.

591. BEHRENS, O.: Der Einfluss der Korrosion auf die Biegeschwingungsfestigkeit von Stählen und Reinnickel (The Influence of Corrosion on the Endurance Limit in Bending of Steels and Pure Nickel), *Mitt. d. Wöhler-Inst., Braunschweig*, No. 15, NEM Verlag, Berlin, 1933, 73 pp.

592. BENGOUGH, G. D.: Corrosion of Metals in Salt Solutions and Seawater, *J. Soc. Chem. Ind.*, v. 52, 1933, pp. 195–210, 228–239.

593. BOEGEHOLD, A. L.: Wear Tests and Value of Hardness Testing for Control of Product, *Trans. Am. Foundrymen's Assoc.*, v. 41, 1933,

594. Briggs, C. W., and R. A. Gezelius: Studies on Solidification and Contraction and Their Relation to the Formation of Hot Tears in Steel Castings, *Trans. Am. Foundrymen's Assoc.*, v. 41, 1933, pp. 385–448.

595. Brophy, G. R., and L. L. Wyman: Factors Influencing the Annealing of Cold-rolled Strip Steel, *Trans. Am. Soc. Steel Treat.*, v. 21, 1933, pp. 532–556.

596. Bühler, H., and H. Buchholtz: Die Wirkung von Eigenspannungen auf die Biegeschwingungsfestigkeit (Influence of Internal Stresses on Endurance), *Stahl u. Eisen*, v. 53, 1933, pp. 1330–1332.

597. Comstock, G. F.: Titanium in Gray Cast Iron, *Iron Age*, v. 131, 1933, pp. 857–859, adv. p. 12.

598. Coyle, F. B.: Making, Drawing, and Forging Dies from Gray Cast Iron, *Steel*, v. 92, June 19, 1933, pp. 26–27.

599. Curran, J. J., and E. A. Sanford: Wrought Iron—Fact and Opinion, *Metals & Alloys*, v. 4, 1933, pp. 1–5.

600. De Haas, W. J., and R. A. Hadfield: On the Effect of the Temperature of Liquid Hydrogen (−252.8°C.) on the Tensile Properties of Forty-one Specimens of Metals Comprising (*a*) Pure Iron 99.85%; (*b*) Four Carbon Steels; (*c*) Thirty Alloy Steels; (*d*) Copper and Nickel; (*e*) Four Non-ferrous Alloys, *Phil. Trans. Royal Soc.*, London, ser. A, v. 232, 1933, pp. 297–332.

601. Delbart, G.: L'Acier moulé—Structure—Traitements thermiques—Propriétés (Cast Steel, Structure, Thermal Treatment, Properties), *Bull. Assoc. Tech. de Fonderie*, v. 7, 1933, pp. 87–110.

602. Desch, C. H.: Recent Developments in Metallurgy—Aging and Age Hardening, *Foundry Tr. J.*, v. 48, 1933, p. 370.

603. Detroit Electric Furnace Company: Effect of Temperature and Stirring on Cast Iron, mimeographed report, March 16, 1933, 8 pp.

604. Donaldson, J. W.: The Thermal Conductivity of Wrought Iron, Steel, Malleable Cast Iron, and Cast Iron, *J. Iron Steel Inst.*, v. 128, 1933, pp. 255–276.

605. Dusold, T.: Der Einfluss der Korrosion auf die Drehschwingungsfestigkeit von Stählen und Nichteisenmetallen (The Influence of Corrosion on the Torsional Endurance Limit of Steels and Nonferrous Metals), *Mitt. d. Wöhler-Inst., Braunschweig*, No. 14, NEM Verlag, Berlin, 1933, 89 pp.

606. Eaton, G. M.: Practical Plasticity Problems, *Mech. Eng.*, v. 55, 1933, pp. 557–560; discussion, pp. 647–649.

607. Egan, J. J., W. Crafts, and A. B. Kinzel: Low-temperature Impact Strength of Some Normalized Low-alloy Steels, *Trans. Am. Soc. Steel Treat.*, v. 21, 1933, pp. 1136–1152. Also: *Metal Progress*, v. 24, No. 3, 1933, pp. 18–20.

608. Esser, H., and G. Müller: Die Gitterkonstanten von reinem Eisen und Eisen-Kohlenstoff-Legierungen bei Temperaturen bis 1100° (Lattice Constants of Pure Iron and Iron-carbon Alloys at Temperatures up to 1100°C.), *Archiv f. d. Eisenhüttenwesen*, v. 7, 1933, pp. 265–268.

609. FAULHABER, R., H. BUCHHOLTZ, and E. H. SCHULZ: Einfluss des Probendurchmessers auf die Biegeschwingungsfestigkeit von Stahl (Influence of the Diameter of the Test Specimen on the Endurance Limit of Steel), *Stahl u. Eisen*, v. 53, 1933, pp. 1106–1108.

610. FOLEY, F. B.: Creep Testing and High-creep Steels, *Iron Age*, v. 132, Dec. 21, 1933, pp. 21–23.

611. FORSMAN, O.: Bidrag till kännedomen om ståls hållfasthet mot utmattning och slag vid låg temperatur (Fatigue and Impact Strength of Steel at Low Temperatures), *Jernk. Ann.*, v. 117 (88), 1933, pp. 519–529.

612. FRENCH, H. J.: Fatigue and the Hardening of Steels, *Trans. Am. Soc. Steel Treat.*, v. 21, 1933, pp. 899–946.

613. GALIBOURG, J.: Sur le vieillissement des métaux après écrouissage par traction (Aging of Metals after Cold Working in Tension), *Rev. de Mét.*, *Mém.*, v. 30, 1933, pp. 96–111.

614. GILLETT, H. W., and H. C. CROSS: Obtaining Reliable Values for Creep of Metals at High Temperatures, *Metals & Alloys*, v. 4, 1933, pp. 91–98, 104.

615. GOUGH, H. J., and D. G. SOPWITH: Some Comparative Corrosion-fatigue Tests Employing Two Types of Stressing Action, *J. Iron Steel Inst.*, v. 127, 1933, pp. 301–335.

616. GRANJON, R.: À propos des déformations soudure et oxycoupage (Distortion in Welding and Oxygen Cutting), *Soudure et Oxycoupage* (Suppl. to *Rev. de la Soudure Autogène*), v. 10, 1933, p. 10.

617. GREGG, A. W.: New Heat Treatment for Carbon Steel Locomotive Castings, *Iron Age*, v. 131, 1933, pp. 192–193, adv. p. 18.

618. GREINER, E. S., J. S. MARSH, and B. STOUGHTON: "The Alloys of Iron and Silicon," McGraw-Hill Book Company, Inc., New York, 1933, 457 pp.

619. GROTTS, F.: Dynamic Properties of Steel Castings, *Trans. Am. Foundrymen's Assoc.*, v. 41, 1933, pp. 538–557.

620. HALL, R.: High-strength and Wear-resistant Malleable Cast Iron, *Trans. Am. Foundrymen's Assoc.*, v. 41, 1933, pp. 112–126.

621. HARDER, O. E., and H. A. GROVE: Hot Hardness of High-speed Steels and Related Alloys, Am. Inst. Min. Met. Eng., *Tech. Pub.* 497, 1933, 39 pp.

622. HENSEL, F. R., and T. F. HENGSTENBERG: Some Observations on the Effects of Inclusions on the Tensile and Dynamic Properties of Certain Ferrous Metals, Am. Inst. Min. Met. Eng., *Preprint*, 1933, 27 pp.

623. Iron and Steel Institute: First Report of the Steel Castings Research Committee, *Special Rep.* 3, The Institute, London, 1933, 56 pp.

624. JOHNSON, J. B., and T. OBERG: Mechanical Properties at Minus 40 Degrees of Metals Used in Aircraft Construction, *Metals & Alloys*, v. 4, 1933, pp. 25–30.

625. KAWAI, T.: Serrated Discontinuity on Load-extension Diagram and Age Hardening of Metals and Alloys, *Sci. Rep.*, Sendai, ser. 1, v. 22, 1933, pp. 354–374.

626. Keulegan, G. H., and M. R. Houseman: Temperature Coefficient of the Moduli of Metals and Alloys Used as Elastic Elements, *Bur. Stand. J. Research*, v. 10, 1933, pp. 289–320.

627. Kinský, F: Formation of Columnar Crystals in Steel Sheets after Normalization, *J. Iron Steel Inst.*, v. 127, 1933, pp. 209–228.

628. Koch, A., and E. Piwowarsky: Einfluss des Kohlenstoffgehaltes auf das Gefüge und die Festigkeitseigenschaften des grauen Gusseisens, unter Berücksichtigung verschiedener Siliziumgehalte, Giesstemperaturen und Wandstärken (Influence of the Carbon Content on the Structure and the Mechanical Properties of Gray Cast Iron, with Consideration of Different Silicon Contents, Casting Temperatures, and Wall Thicknesses), *Giesserei*, v. 20, 1933, pp. 1–7, 26–31.

629. Krynitsky, A. I.: Surface Tension of Molten Metals, *Metals & Alloys*. v. 4, 1933, pp. 79–84.

630. Leitner, F.: Der Einfluss des Grundwerkstoffes auf die Bildung von Fehlstellen bei der Lichtbogenschweissung (The Influence of the Base Metal on the Formation of Defects in Electric-arc Welding), *Archiv f. d. Eisenhüttenwesen*, v. 7, 1933, pp. 311–314.

631. Lequis, W., H. Buchholtz, and E. H. Schulz: Biegeschwingungsfestigkeit und Kerbempfindlichkeit in ihrer Beziehung zu den übrigen Festigkeitseigenschaften bei Stahl (Relation of Fatigue Strength and Notch Sensitiveness to the Other Strength Properties of Steel), *Stahl u. Eisen*, v. 53, 1933, pp. 1133–1137.

632. Logan, K. H.: The Protection of Pipes against Soil Action, *Trans. Electrochem. Soc.*, v. 64, 1933, pp. 113–124.

633. Luerssen, G. V., and O. V. Greene: The Torsion Impact Test, *Proc. Am. Soc. Test. Mat.*, v. 33, part 2, 1933, pp. 315–333.

634. Lyse, I., and H. J. Godfrey: Shearing Properties and Poisson's Ratio of Structural and Alloy Steels, *Proc. Am. Soc. Test. Mat.*, v. 33, part 2, 1933, pp. 274–292.

635. Mailänder, R.: Über die Dauerfestigkeit von nitrierten Proben (The Fatigue of Nitrided Specimens), *Z. Ver. deutsch. Ing.*, v. 77, 1933, pp. 271–274. Also: *Tech. Mitt. Krupp*, July, 1933, No. 2, pp. 53–58.

636. Maurer, E., and H. Korschan: Beitrag zur Kenntnis der mechanischen Eigenschaften grösster Schmiedestücke (Mechanical Properties of Large Forgings), *Stahl u. Eisen*, v. 53, 1933, pp. 209–215, 243–251.

637. McMullan, O. W.: Martensitic Grains in Air-cooled Low-carbon Steel and Their Effect on Machinability, *Trans. Am. Soc. Steel Treat.*, v. 21, 1933, pp. 1034–1060.

638. Mears, R. B.: Combating Rust with Metallic Finishes, *Bell Lab. Record*, v. 11, 1933, pp. 141–144.

639. Merz, A., and H. Schuster: Über die Graphitisierung von weissem Gusseisen (The Graphitization of White Cast Iron), *Giesserei*, v. 20, 1933, pp. 145–151, 173–181.

640. Mitinsky, A.: Le module d'élasticité des métaux à différentes températures (The Elastic Modulus of Metals at Different Temperatures), *Chimie et Industrie*, Special No., v. 29, 1933, pp. 616–619.

641. MOORE, H. F.: Corrosion Fatigue of Metals, *Metals & Alloys*, v. 4, 1933, pp. 39–40.

642. MOORE, H. F., and H. B. WISHART: An "Overnight" Test for Determining Endurance Limit, *Proc. Am. Soc. Test. Mat.*, v. 33, part 2, 1933, pp. 334–347.

643. MURPHY, D. W.: Behavior of Steel in Sulphur-containing Atmospheres at Forging Temperatures, *Trans. Am. Soc. Steel Treat.*, v. 21, 1933, pp. 510–531.

644. National Bureau of Standards: Standard and Specifications for Metals and Metal Products, *Misc. Publ.* No. 120, 1933, pp. 706–742.

645. NEUSTÄTTER, K.: Beitrag zur Kenntnis des Kupolofenbetriebs beim Erschmelzen von hochwertigem grauen Gusseisen (Cupola Furnace Practice in Melting High-test Gray Iron), *Mitt. Forsch.-Anst. d. G.H.H. Konzerns*, v. 2, 1933, pp. 112–134.

646. PALMER, F. R.: Factors That Affect Machinability, *Metal Progress*, v. 23, No. 4, 1933, pp. 17–19.

647. PÉRIN, S.: Contribution à l'étude de la coulabilité des fontes (Castability of Irons), *Fonderie Belge*, v. 3, 1933, pp. 160–175. Also (présenté par R. Berger): *Bull. Assoc. Tech. de Fonderie*, v. 7, 1933, pp. 489–506. Abst.: *Metals & Alloys*, v. 4, 1933, p. 176. Abst.: *Rev. de Mét., Mém.*, v. 31, 1934, pp. 377–380.

648. PIWOWARSKY, E.: Die Auflösungsgeschwindigkeit des Graphits in flüssigem Eisen (Rate of Solution of Graphite in Molten Iron), *Archiv f. d. Eisenhüttenwesen*, v. 7, 1933, pp. 431–432.

649. PIWOWARSKY, E., B. BOŽIĆ, and E. SÖHNCHEN: Zugfestigkeit und Einschnürung von Stahlguss bei 650 bis 1450° (Tensile Strength and Reduction of Area of Cast Steels at Temperatures between 650 and 1450°C.), *Archiv f. d. Eisenhüttenwesen*, v. 7, 1933, pp. 127–130.

650. PIWOWARSKY, E., and E. SÖHNCHEN: Wandstärkenempfindlichkeit und Treffsicherheit bei unlegiertem und legiertem Gusseisen (Wall-thickness Susceptibility and Reliability of Unalloyed and Alloyed Cast Irons), *Z. Ver. deutsch. Ing.*, v. 77, 1933, pp. 463–468. Abst.: *Stahl u. Eisen*, v. 54, 1934, pp. 267–268.

651. POHL, W. M.: Vibrations of Structures and Materials, *Product Engineering*, v. 4, 1933, pp. 91–94.

652. POMP, A., and B. ZAPP: Beitrag zur akustischen Werkstoffuntersuchung von Stahlstäben insbesondere für die Herstellung von Gongspielen (Acoustic Testing of Steel Rods Especially for the Manufacture of Gongs), *Mitt. K.-W. Inst. Eisenforschung*, v. 15, 1933, pp. 21–35.

653. SAEGER, C. M., JR., and E. J. ASH: Properties of Gray Cast Iron as Affected by Casting Conditions, *Trans. Am. Foundrymen's Assoc.*, v. 41, 1933, pp. 449–468.

654. SCHMIDT, W.: Über thermische Ausdehnungsmessungen an Eisen mittels Röntgenstrahlen (Thermal Expansion Measurements on Iron by Means of X-rays), "Ergebnisse der technischen Röntgenkunde," Akademische Verlagsgesellschaft m. b. H., Leipzig, 1933, v. 3, pp. 194–201.

655. Schwarz, M. von: Gusseisen-Gefügelehre (The Structure of Cast Iron), *Forschungsarb. über Metallkunde u. Röntgenmetallographie*, No. 10, 1933, 164 pp.

656. Schwarz, M. von, and A. Väth: Beziehungen zwischen der Wandstärke, dem Stabdurchmesser und dem Maurer-Diagramm für Gusseisen (Relations between Wall Thickness, Specimen Diameter, and the Maurer Diagram for Cast Iron), *Giesserei*, v. 20, 1933, pp. 373–376.

657. Scorah, R. L.: Models Reveal Changes Needed to Combat Fatigue, *Machine Design*, v. 5, November, 1933, pp. 31–33.

658. Shelton, S. M., and W. H. Swanger: Thermal Conductivity of Irons and Steels and Some Other Metals in the Temperature Range 0 to 600°C., *Trans. Am. Soc. Steel Treat.*, v. 21, 1933, pp. 1061–1078.

659. Simonds, H. R.: Permanent Molds in Manufacturing Cream Separator Parts, *Iron Age*, v. 131, 1933, pp. 940–943.

660. Söhnchen, E., and E. Piwowarsky: Einfluss der chemischen Zusammensetzung von Gusseisen auf seinen Verschleiss (Influence of the Chemical Composition of Cast Iron on Its Wear), *Archiv f. d. Eisenhüttenwesen*, v. 7, 1933, pp. 371–372.

661. Sommer, F.: Physikalische Eigenschaften von Gasschmelz- und Lichtbogen-Schweissen bei Baustählen höherer Festigkeit (Physical Properties of Gas and Electric Welds in High-strength Structural Steels), *Autogene Metallbearbeitung*, v. 26, 1933, pp. 22–27.

662. Suslov, B. M.: Rail Failures in Siberian Winters, *Metal Progress*, v. 24, No. 6, 1933, pp. 45–47.

663. Tapsell, H. J.: Creep of Steels, Soc. Chem. Ind., Chem. Eng. Group, February, 1933, 8 pp. Abst.: *Mech. World and Eng. Record*, v. 93, 1933, pp. 236–237, 263–265, 312–314.

664. Templin, R. L.: The Fatigue Properties of Light Metals and Alloys, *Proc. Am. Soc. Test. Mat.*, v. 33, part 2, 1933, pp. 364–386.

665. Thum, E. E. (editor): "The Book of Stainless Steels," American Society for Steel Treating, Cleveland, 1933, 631 pp.

666. Tiemann, H. P.: "Iron and Steel," McGraw-Hill Book Company, Inc., New York, 3d ed., 1933, 590 pp.

667. Uhlitzsch, H., and W. Weichelt: (The Maurer Cast Iron Diagram), Dissertation, Sächs. Bergakademie Freiberg, 1933. Abst.: *Stahl u. Eisen*, v. 54, 1934, p. 1091.

668. Upthegrove, C.: Scaling at Heat Treating Temperatures, *Metal Progress*, v. 23, No. 6, 1933, pp. 30–33, 62.

669. Upthegrove, C., and D. W. Murphy: Scaling of Steel at Heat Treating Temperatures, *Trans. Am. Soc. Steel Treat.*, v. 21, 1933, pp. 73–96.

670. Verein deutscher Ingenieure, Fachausschuss für Maschinenelemente (Committee on the Principles of Machine Design): Dauerfestigkeits-Schaubilder (Endurance-limit Diagrams), *Z. Ver. deutsch. Ing.*, v. 77, 1933, suppl. to No. 42, Oct. 21, 4 pp., and: Dauerfestigkeits-Schaubilder von Kohlenstoff-Stählen (Endurance-limit Diagrams for Carbon Steels), *ibid.*, suppl. to No. 50, Dec. 16, 2 pp.

671. WHITE, A. E., and R. SCHNEIDEWIND: Effect of Superheat on Annealing of Malleable Iron, *Trans. Am. Foundrymen's Assoc.*, v. 41, 1933, pp. 98–111.
672. WOOD, T. J.: Growth of Gray Cast Irons, *Iron Age*, v. 132, July 13, 1933, pp. 12–13.
673. WOOD, T. J.: Growth of Cast Irons at High Temperatures, *Iron Age*, v. 132, July 27, 1933, pp. 12–14.
674. WOOD, T. J.: Castings That Keep Strength and Shape at High Temperatures, *Iron Age*, v. 132, Aug. 3, 1933, pp. 30–32.

1934

675. ABEGG, R., F. AUERBACH, and I. KOPPEL (editors): "Handbuch der anorganischen Chemie" (Handbook of Inorganic Chemistry), S. Hirzel, Leipzig, v. 4, section 3, part 2A, No. 4, 1934.
676. ABRAM, H. H.: The Value of Young's Modulus for Steel, *J. Iron Steel Inst.*, v. 129, 1934, pp. 325–330.
677. American Society for Testing Materials: Report of Special Subcommittee (of Committee A-3) on Test Bars, *Proc. Am. Soc. Test. Mat.*, v. 34, part 1, 1934, pp. 148–153.
678. ANON.: Failure of Rail Steel, *Iron Age*, v. 133, Feb. 8, 1934, p. 41-B.
679. AUSTIN, J. B., and R. H. H. PIERCE, JR.: The Linear Thermal Expansion and α-γ Transformation Temperature (A_3 Point) of Pure Iron, *Trans. Am. Soc. Metals*, v. 22, 1934, pp. 447–470.
680. BOLTON, J. W.: Heat Treatment of Cast Iron, *Foundry*, v. 62, 1934, No. 8, pp. 18–20, 48; No. 9, pp. 16–18, 56, 59.
681. British Cast Iron Research Association: Thirteenth Annual Report for the Year Ending June 30, 1934, The Association, London, 1934, 23 pp.
682. CROSS, H. C.: High-temperature Tensile, Creep, and Fatigue of Cast and Wrought High- and Low-carbon, 18 Cr 8 Ni Steel from Split Heats, *Trans. Am. Soc. Mech. Eng.*, v. 56, 1934, pp. 533–553.
683. DAUVERGNE, J.: Les différentes qualités d'acier doux envisagées du point de vue de leur résistance à chaud (Different Qualities of Soft Steels from the Standpoint of Their Resistance to Heat), *Chimie et Industrie*, Special No., April, 1934, pp. 609–613.
684. DAVENPORT, E. S., E. L. ROFF, and E. C. BAIN: Microscopic Cracks in Hardened Steel, Their Effects and Elimination, *Trans. Am. Soc. Metals*, v. 22, 1934, pp. 289–310.
685. DEJEAN, P., and S. GERSZONOWICZ: "État actuel de l'essai de fragilité des métaux" (Present Status of Testing Metals for Brittleness), Dunod, Paris, 1934, 103 pp.
686. DENISON, I. A., and R. B. HOBBS: Corrosion of Ferrous Metals in Acid Soils, *Bur. Stand. J. Research*, v. 13, 1934, pp. 125–150.
687. DIEPSCHLAG, E.: Die Gusshaut (The Casting Skin), *Giesserei*, v. 21, 1934, pp. 408–412.
688. DONALDSON, J. W.: The Thermal and Electrical Properties of Cast Iron, *Foundry Tr. J.*, v. 50, 1934, pp. 283–285.
689. DOWLING, J. J., S. M. DIXON, and M. A. HOGAN: Fatigue Tests on Hard Drawn Steel Wire, *Engineer*, v. 157, 1934, pp. 424–426.

690. Dustin, H.: (Report of) Belgian Research Committee on the Behavior of Metals at Elevated Temperatures, *J. Iron Steel Inst.*, v. 130, 1934, pp. 127–141.

691. Eakin, C. T.: A Review of Spring Wire Characteristics, *Iron Age*, v. 134, Aug. 16, 1934, pp. 16–19, 76, 78.

692. Eilender, W. W. Oertel, and H. Schmalz: Grundsätzliche Untersuchungen des Verschleisses auf der Spindel-Maschine (Fundamental Studies of Wear with the Spindel Machine), *Archiv f. d. Eisenhüttenwesen*, v. 8, 1934, pp. 61–65.

693. Eisenkolb, F.: Über die Bewitterungsversuche mit Feinblechen aus weichem Stahl (Weather-exposure Tests with Soft Steel Sheets), *Korrosion u. Metallschutz*, v. 10, 1934, pp. 161–166.

694. Enders, W.: Einfluss der Gefügeausbildung auf die Dauerstandfestigkeit des Stahles (Influence of Structure on the Creep of Steels), *Mitt. K.-W. Inst. Eisenforschung*, v. 16, 1934, pp. 159–167.

695. Epstein, S., J. H. Nead, and T. S. Washburn: Grain-size Control of Open-hearth Carbon Steels, *Trans. Am. Soc. Metals*, v. 22, 1934, pp. 942–978.

696. Esser, H., V. Greis, and W. Bungardt: Über die innere Reibung von flüssigem Roheisen (Inner Friction in Liquid Pig Iron), *Archiv f. d. Eisenhüttenwesen*, v. 7, 1934, pp. 385–388.

697. Esser, H., and G. Ostermann: Einfluss der Wärmebehandlung auf die magnetische Sättigung unlegierter Stähle (Influence of Heat Treatment on Magnetic Saturation of Unalloyed Steels), *Archiv f. d. Eisenhüttenwesen*, v. 8, 1934, pp. 173–176.

698. Finkeldey, W. H.: The Early Interpretation of Test Results in the Atmospheric Corrosion of Nonferrous Metals and Alloys, "Symposium on the Outdoor Weathering of Metals and Metallic Coatings," Washington Regional Meeting, American Society for Testing Materials, March 7, 1934, pp. 69–98.

699. Föppl, O.: Die Qualifikation der Werkstoffe mit Hilfe der Werkstoffdämpfung (Appraisal of Materials by Damping Properties), *Mitt. d. Wöhler-Inst., Braunschweig*, No. 18, NEM Verlag, Berlin, 1934, 8 pp.

700. French, H. J., and J. W. Sands (editors): "Nickel Alloy Steels," International Nickel Co., Inc., New York, 1934, 7 sections.

701. Gill, E. T., and R. Goodacre: Some Aspects of the Fatigue Properties of Patented Steel Wire, *J. Iron Steel Inst.*, v. 130, 1934, pp. 293–323.

702. Gillett, H. W.: Properties of Ferrite as Revealed by Scratch Hardness Tests, *Metals & Alloys*, v. 5, 1934, pp. 159–160. Extended abstract of theses by Briggs and Williams, Stanford University.

703. Gould, A. J.: The Effects of Superposing Cyclic Stressing upon Chemical Corrosion, *Engineering*, v. 138, 1934, pp. 79–81.

704. Graham, H. W.: Is There a Test for Machinability? *Metals & Alloys*, v. 5, 1934, pp. 93–95.

705. Gregg, J. L., and B. N. Daniloff: "The Alloys of Iron and Copper," McGraw-Hill Book Company, Inc., New York, 1934, 454 pp.

706. Griffiths, R.: The Blistering of Iron Oxide Scales and the Conditions for the Formation of a Non-adherent Scale, *J. Iron Steel Inst.*, v. 130, 1934, pp. 377–388.

707. GRÜN, P.: Über die im abgekürzten Verfahren ermittelte Dauerstandfestigkeit von Stählen in Abhängigkeit von verschiedenen Legierungszusätzen und von der Wärmebehandlung, sowie ein Beitrag zur Frage des Dehnverhaltens niedriglegierter Stähle (Short-time Tests on Creep of Steels in Relation to Alloying and Heat Treatment, and a Discussion of the Elastic Behavior of Low-alloy Steels), *Mitt. Forsch.-Inst. Vereinigte Stahlwerke, A. G.*, v. 4, 1934, pp. 113–160. Condensed: *Archiv f. d. Eisenhüttenwesen*, v. 8, 1934, pp. 205–211.

708. GRUSCHKA, G.: Zugfestigkeit von Stählen bei tiefen Temperaturen (Tensile Strength of Steels at Low Temperatures), *Forschungsheft* 364, suppl. to *Forsch. a. d. Gebiete d. Ingenieurwesens*, Series B, No. 5, VDI Verlag, Berlin, 1934, 20 pp.

709. HATFIELD, W. H.: Steel Castings, *Trans. Am. Foundrymen's Assoc.*, v. 42, 1934, pp. 672–704.

710. HATTORI, D.: The Thermal Conductivity of Tool Steels, *J. Iron Steel Inst.*, v. 129, 1934, pp. 289–309.

711. HAYES, A., and R. O. GRIFFIS: Non-aging Iron and Steel for Deep Drawing, *Metals & Alloys*, v. 5, 1934, pp. 110–112.

712. HEIMKE, H. W., and W. C. SCHULTE: Low-temperature Impact Tests of Medium-manganese Steel Plate, *Metals & Alloys*, v. 5, 1934, pp. 31–36.

713. HEINDLHOFER, K.: The "Plasticity" of Iron at Low Temperatures, *Metals Technology*, v. 1, No. 7, 1934, 7 pp. Also: *Trans. Am. Inst. Min. Met. Eng.*, v. 116, 1935, pp. 232–238.

714. HEINDLHOFER, K., and B. M. LARSEN: Why Do Some Metals Resist Oxidation?, *Metal Progress*, v. 26, No. 3, 1934, pp. 34–37.

715. HELLER, P. A., and H. JUNGBLUTH: Die Wandstärken-Empfindlichkeit getrennt gegossener Probestäbe und ihre Beziehung zur Analyse (Size Effect in Individually Cast Test Bars and Its Relation to Analysis), *Tech. Mitt. Krupp*, v. 2, 1934, pp. 106–116. Condensed: *Archiv f. d. Eisenhüttenwesen*, v. 8, 1934, pp. 75–82.

716. HEROLD, W.: "Wechselfestigkeit metallischer Werkstoffe" (Endurance of Metals), Julius Springer, Wien, 1934, 276 pp.

717. HERTY, C. H., JR., and D. L. MCBRIDE: Effect of Deoxidation on the Impact Strength of Carbon Steels at Low Temperatures, *Cooperative Bull.* 67, Min. Met. Advisory Boards to Carnegie Inst. Technology and Bur. Mines, Pittsburgh, 1934, 43 pp.

718. HIPPENSTEEL, C. L.: Galvanic Corrosion by Contact of Dissimilar Metals, "Symposium on the Outdoor Weathering of Metals and Metallic Coatings," Washington Regional Meeting, American Society for Testing Materials, March 7, 1934, pp. 99–109.

719. HOCKER, C. D.: Outdoor Test Results on Bare and Metal-coated Ferrous Specimens, "Symposium on the Outdoor Weathering of Metals and Metallic Coatings," Washington Regional Meeting, American Society for Testing Materials, March 7, 1934, pp. 1–27.

720. HOPKINS, R. K.: Impact Resistance of Some Steels and Welds at Sub-zero Temperatures, *J. Am. Welding Soc.*, v. 13, No. 10, 1934, pp. 16–28.

721. Hughes, T. P., and R. L. Dowdell: Quenching Steels in Hot Lead, *Trans. Am. Soc. Metals*, v. 22, 1934, pp. 737–750.

722. Hurst, J. E.: Permanent Set and Elastic Hysteresis in Cast Iron, *Foundry Tr. J.*, v. 51, 1934, pp. 405–406; v. 52, 1935, pp. 63–64, 66.

723. Institution of Automobile Engineers: Cylinder Wear, *Engineer*, v. 155, 1934, pp. 634–636, 660–662.

724. Johnston, J.: Corrosion Problems, *Ind. Eng. Chem.*, v. 26, 1934, pp. 1238–1244.

725. Keil, O. von, R. Mitsche, A. Legat, and M. Trenkler: Der Einfluss nichtmetallischer Keime auf die Graphitausbildung im Gusseisen. Vererbungserscheinungen im Gusseisen und deren Ursachen (The Influence of Nonmetallic Nuclei on Graphite Formation in Cast Iron. Heredity Phenomena in Cast Iron and Their Causes), *Archiv f. d. Eisenhüttenwesen*, v. 7, 1934, pp. 579–584.

726. Kenyon, R. L., and R. S. Burns: Testing Sheets for Blue Brittleness and Stability against Changes Due to Aging, *Proc. Am. Soc. Test. Mat.*, v. 34, part 2, 1934, pp. 48–58.

727. Knipp, E.: Alterung und Warmzugfestigkeit von Stahlguss (Aging and High-temperature Tensile Strength of Steel Castings), *Stahl u. Eisen*, v. 54, 1934, pp. 777–778.

728. Körber, F.: Qualités de résistance des aciers à chaud (Tensile Properties of Steel at Elevated Temperatures), *Rev. de Mét.*, *Mém.*, v. 31, 1934, pp. 359–368.

729. Köster, W., and W. Tonn: Zusammenhang zwischen dem Gefügeaufbau der Eisenlegierungen und ihrem Verschleiss bei der Prüfung nach M. Spindel (Relation between the Structure of Iron Alloys and Their Wear in the Spindel Test), *Archiv f. d. Eisenhüttenwesen*, v. 8, 1934, pp. 111–115.

730. Krouse, G. N.: A High-speed Fatigue Testing Machine and Some Tests of Speed Effect on Endurance Limit, *Proc. Am. Soc. Test. Mat.*, v. 34, part 2, 1934, pp. 156–164.

731. Lake, E. F.: Variations in Properties when Treating Large Masses, *Heat Treating Forging*, v. 20, 1934, pp. 128–130.

732. Lemoine, R. P.: Cupola High-test Cast Iron, *Trans. Am. Foundrymen's Assoc.*, v. 42, 1934, pp. 745–761.

733. Lessells, J. M.: Why the Elastic Modulus of Iron Is Not a Constant, *Product Engineering*, v. 5, 1934, pp. 453–454.

734. Logan, K. H., and R. H. Taylor: Soil-corrosion Studies, 1932. Rates of Loss of Weight and Pitting of Ferrous and Non-ferrous Specimens and Metallic Protective Coatings, *Bur. Stand. J. Research*, v. 12, 1934, pp. 119–145.

735. Lorig, C. H., and C. S. Smith: Effect of Copper in Malleable Cast Iron, *Trans. Am. Foundrymen's Assoc.*, v. 42, 1934, pp. 211–236.

736. Ludewig, J. W.: Torsional Moduli Variations of Spring Materials with Temperature, *Trans. Am. Soc. Metals*, v. 22, 1934, pp. 833–860.

737. MacKenzie, J. T.: The Meaning of the Transverse Test of Cast Iron to the Designing Engineer, *Foundry Tr. J.*, v. 51, 1934, pp. 129–132.

738. MAURER, E., and H. GUMMERT: Gefüge und Festigkeitseigenschaften grosser Schmiedestücke (Structure and Properties of Large Forgings), *Stahl u. Eisen*, v. 54, 1934, pp. 1281–1289, 1309–1320.

739. MCADAM, D. J., JR., and R. W. CLYNE: Influence of Chemically and Mechanically Formed Notches on Fatigue of Metals, *Bur. Stand. J. Research*, v. 13, 1934, pp. 527–572.

740. MCVETTY, P. G.: The Interpretation of Creep Tests, *Proc. Am. Soc. Test. Mat.*, v. 34, part 2, 1934, pp. 105–122. Abst.: *Iron Age*, v. 134, July 26, 1934, pp. 16–19.

741. MELLOR, J. W.: "A Comprehensive Treatise on Inorganic and Theoretical Chemistry," Longmans, Green & Company, London, 1934, v. 13, 948 pp.

742. MOORE, H. F., and J. J. PICCO: Fatigue Tests of High-strength Cast Irons, *Trans. Am. Foundrymen's Assoc.*, v. 42, 1934, pp. 525–542.

743. MUTCHLER, W.: The Weathering of Aluminum Alloy Sheet Materials Used in Aircraft, National Advisory Committee for Aeronautics, *Rep.* 490, 1934, 35 pp.

744. NIPPER, H.: Contribution to the Study of Graphite Formation and Structure in Cast Iron and Its Influence on the Properties of the Cast Metal, *Foundry Tr. J.*, v. 51, 1934, pp. 7–14, 41–42, 52.

745. O'NEILL, H.: "The Hardness of Metals and Its Measurement," Chapman and Hall, Ltd., London, 1934, 292 pp.

746. PARTRIDGE, E. P., and W. C. SCHROEDER: A Critical Summary of Published Information Relating to the Embrittlement of Boiler Steel, unpublished report to Subcommittee of the Joint Research Committee for Boiler Feed Studies, Dec. 1, 1934, mimeographed, 52 pp. (presented at A.S.M.E. meeting).

747. PASCHKE, M., and H. SCHUSTER: Versuche über den Einfluss von Zusätzen eines hochgekohlten Spezialroheisens auf die Wachstums- und Zunderbeständigkeit von Grauguss (Influence of Additions to a High-carbon Special Pig Iron on Growth and Scale Resistance of Gray Iron), *Giesserei*, v. 21, 1934, pp. 469–476.

748. PASSANO, R. F.: The Harmony of Outdoor Weathering Tests, "Symposium on the Outdoor Weathering of Metals and Metallic Coatings," Washington Regional Meeting, American Society for Testing Materials, March 7, 1934, pp. 28–50.

749. PEARCE, J. G.: Modern Cast Irons in Chemical Engineering, *J. Soc. Chem. Ind.*, v. 12, 1934, pp. 1052–1058.

750. PFANNENSCHMIDT, C.: Die Schwingungsfestigkeit von Gusseisen und ihre Beziehungen zum Aufbau und zur Oberflächenbeschaffenheit (Endurance of Cast Iron and Its Relations to Structure and Surface Condition), *Giesserei*, v. 21, 1934, pp. 223–228, 243–245.

751. PIWOWARSKY, E.: "Allgemeine Metallkunde" (General Metallurgy), Gebrüder Bornträger, Berlin, 1934, 248 pp.

752. PIWOWARSKY, E.: L'importance des processus de solidification eutectique sur les propriétés de la fonte (Eutectic Solidification of Cast Iron), *Bull. Assoc. Tech. de Fonderie*, v. 8, 1934, pp. 96–105.

753. Pomp, A.: Über grobkörnige Rekristallisation hochkohlenstoffhaltigen Stahles (Coarse-grained Recrystallization of High-carbon Steel), *Mitt. K.-W. Inst. Eisenforschung*, v. 16, 1934, pp. 9–13.

754. Pomp, A.: Einfluss des Kohlenstoffgehaltes und der Patentierungsbedingungen auf die Festigkeitseigenschaften gezogener Stahldrähte (Influence of Carbon Content and of Patenting Practice on the Strength of Cold-drawn Steel Wire), *Mitt. K.-W. Inst. Eisenforschung*, v. 16, 1934, pp. 117–127.

755. Pomp, A., and H. Herzog: Zur Frage der Dauerstandfestigkeit des Stahles oberhalb 500° (The Creep Problem in Steels above 500°C.), *Mitt. K.-W. Inst. Eisenforschung*, v. 16, 1934, pp. 141–153.

756. Price, B. K.: Improves Quality of Centrifugally Cast Pipe, *Steel*, v. 94, Feb. 12, 1934, pp. 23–25.

757. Ranque, G., and P. Henry: La méthode d'autostabilisation thermique et son utilisation à l'étude de quelques aciers résistant à chaud (Method of Thermal Self-stabilization and Its Utilization in the Study of Some Heat-resistant Steels), *Rev. de Mét., Mém.*, v. 31, 1934, pp. 248–265.

758. Roesch, K.: Deutscher Temperguss als Werkstoff im Maschinen- und Apparatebau (German Malleable Iron as Structural Material in Machinery and Apparatus), *Giesserei*, v. 21, 1934, pp. 264–271.

759. Rolf, R. L.: Coarse- and Fine-grained Steels, *Heat Treating Forging*, v. 20, 1934, pp. 271–276, 331–333.

760. Rosenberg, S. J., and L. Jordan: Influence of Oxide Films on the Wear of Steels, *Bur. Stand. J. Research*, v. 13, 1934, pp. 267–280.

761. Ruder, W. E.: The Influence of Grain Size on Magnetic Properties, *Trans. Am. Soc. Metals*, v. 22, 1934, pp. 1120–1141.

762. Sachs, G.: Die Blechprüfung mit dem Erichsengerät (Testing Sheet with the Erichsen Apparatus), *Metallwirtschaft*, v. 13, 1934, pp. 79–81.

763. Sale, P. D.: Compression Tests of Structural Steel at Elevated Temperatures, *Bur. Stand. J. Research*, v. 13, 1934, pp. 713–743.

764. Schane, P., Jr.: Effects of McQuaid-Ehn Grain Size on the Structure and Properties of Steel, *Trans. Am. Soc. Metals*, v. 22, 1934, pp. 1038–1050.

765. Scheil, E., and W. Tonn: Vergleich von Brinell- und Ritzhärte (Relation between Brinell and Scratch Hardness), *Archiv f. d. Eisenhüttenwesen*, v. 8, 1934, pp. 259–262.

766. Schmidt, G.: Einfluss der im technischen Eisen enthaltenen Verunreinigungen auf die Kerbzähigkeit (Influence of Impurities in Commercial Iron on Impact), *Archiv. f. d. Eisenhüttenwesen*, v. 8, 1934, pp. 263–267.

767. Schramm, G. N., and E. S. Taylerson: The Influence of Rainfall and Smoke on the Corrosion of Iron and Steel, "Symposium on the Outdoor Weathering of Metals and Metallic Coatings," Washington Regional Meeting, Am. Soc. Test. Mat., March 7, 1934, pp. 51–68.

768. Schulz, E. H.: A Low-alloy Rail for Arctic Railroads, *Metal Progress*, v. 26, No. 1, 1934, p. 48.

769. Schwinning, W., M. Knoch, and K. Uhlemann: Wechselfestigkeit und Kerbempfindlichkeit der Stähle bei hohen Temperaturen

(Resistance to Alternating Stress and Notch Sensitivity of Steels at Elevated Temperatures), *Z. Ver. deutsch. Ing.*, v. 78, 1934, pp. 1469–1476.

770. Scott, H.: Factors Determining the Impact Resistance of Hardened Carbon Steels, *Trans. Am. Soc. Metals*, v. 22, 1934, pp. 1142–1173.

771. Shelton, S. M.: Thermal Conductivity of Some Irons and Steels over the Temperature Range 100 to 500°C., *Bur. Stand. J. Research*, v. 12, 1934, pp. 441–450.

772. Sims, C. E.: Preparation of Steel to Avoid Porosity in Castings, *Trans. Am. Foundrymen's Assoc.*, v. 42, 1934, pp. 323–338.

773. Sirovich, G., and G. Vanzetti: Investigation of the Composition of Common Cast Iron for Boilers, *Trans. Am. Foundrymen's Assoc.*, v. 42, 1934, pp. 71–93.

774. Söhnchen, E.: Die magnetischen Eigenschaften des Gusseisens (Magnetic Properties of Cast Iron), *Archiv f. d. Eisenhüttenwesen*, v. 8, 1934, pp. 29–36.

775. Söhnchen, E.: Die elektrische und die Wärmeleitfähigkeit von Gusseisen (Electric and Thermal Conductivities of Cast Iron), *Archiv f. d. Eisenhüttenwesen*, v. 8, 1934, pp. 223–229.

776. Speller, F. N.: The Corrosion Problem with Respect to Iron and Steel, *Trans. Am. Inst. Min. Met. Eng.*, v. 113, 1934, pp. 13–33.

777. Stine, W. E.: Lowering Welding Costs by Careful Selection of Steels, *Metals & Alloys*, v. 5, 1934, pp. 74–76.

778. Tammann, G., and G. Bandel: Wärmeinhalt und spezifisches Volumen der Eisen-Kohlenstoff-Legierungen (Heat Capacity and Specific Volume of Iron-carbon Alloys), *Archiv f. d. Eisenhüttenwesen*, v. 7, 1934, pp. 571–578.

779. Thyssen, H.: The Influence of Silicon and Aluminum on the Resistance of Cast Iron to High Temperatures, *J. Iron Steel Inst.*, v. 130, 1934, pp. 153–218.

780. Väth, A.: "Der Schleuderguss" (Centrifugal Castings), VDI Verlag, Berlin, 1934, 107 pp.

781. Walther, H.: Dissipation Constants in Solids, *Bell Lab. Record*, v. 12, 1934, pp. 363–366.

782. Walther, H. H.: Electric Cast Iron Practice, *Foundry Tr. J.*, v. 51, 1934, pp. 145–148.

783. White, A. E., and C. L. Clark: Comparison of Single-step Long-time Creep Results with Hatfield's Time-yield Stress, *Trans. Am. Soc. Metals*, v. 22, 1934, pp. 481–504.

784. White, A. E., and C. L. Clark: Influence of Grain Size on the High-temperature Characteristics of Ferrous and Nonferrous Alloys, *Trans. Am. Soc. Metals*, v. 22, 1934, pp. 1069–1098.

785. White, A. E., C. L. Clark, and R. L. Wilson: Qualifications of Steels for High-temperature Service in Petroleum Refinery Equipment, *Oil and Gas J.*, v. 33, Aug. 2, 1934, pp. 16, 18–19, 51.

786. White, A. E., C. L. Clark, and R. L. Wilson: Factors Influencing Creep of Alloy Steels Used in Refinery Equipment at High Temperature, *Oil and Gas J.*, v. 33, Aug. 9, 1934, pp. 16, 33, 34, 36.

787. WIECKER, H.: Die Biegewechselfestigkeit genuteter Stäbe und die Erhöhung der Dauerhaltbarkeit durch das Oberflächendrücken (Resistance to Repeated Bending of Specimens with Keyways and the Improvement in Endurance through Working the Surface), *Mitt. d. Wöhler-Inst., Braunschweig,* No. 19, NEM Verlag, Berlin, 1934, 52 pp.

788. WILSON, R. L.: Grain Size in Steel, *Metal Progress,* v. 26, No. 2, 1934, pp. 17–22.

1935

789. ANON.: Effect of Superheating on Cast Iron Test Bars, *Bur. Stand. Tech. News Bull.* 213, January, 1935, p. 6.

790. ARMSTRONG, T. N.: Properties of Some Cast Alloy Steels, *Trans. Am. Soc. Metals,* v. 23, 1935, pp. 286–318.

791. Bethlehem Steel Company: "Bethlehem Alloy Steels" (Catalog 107), The Company, Bethlehem, 1935, 375 pp.

792. BONZEL, M.: "Le tréfilage de l'acier" (Wire Drawing), Dunod, Paris, 1934, 638 pp. (Translated into English by K. B. Lewis: "Steel Wire, Manufacture and Properties," Engineers Book Shop, New York, 1935, 495 pp.)

793. CLARK, P. H., and E. L. ROBINSON: An Automatic Creep Test Furnace Guide, *Metals & Alloys,* v. 6, 1935, pp. 46–49.

794. CLEAVES, H. E., and J. G. THOMPSON: "The Metal—Iron," McGraw-Hill Book Company, Inc., New York, 1935, 574 pp.

795. DAVENPORT, E. S., and E. C. BAIN: The Aging of Steel, *Trans. Am. Soc. Metals,* v. 23, 1935, pp. 1047–1106.

796. ENDO, K.: On the Oxidation of Iron and Steel at Elevated Temperature and the Structure of the Scale, Reports of Research Laboratory, Nippon-Seitetsu-Yawata-Steel-Works, v. 14, Jan. 30, 1935, pp. 1–16. (In Japanese.) Abst.: *Metals & Alloys,* v. 6, 1935, p. MA 374L5.

797. GILL, E. T., and R. GOODACRE: Some Aspects of the Fatigue Properties of Patented Steel Wires. II. Note on the Effect of Low-temperature Heat Treatment, *J. Iron Steel Inst.,* v. 132, 1935, pp. 143–177.

798. GILLETT, H. W.: Phosphorus as an Alloying Element in Steel, II, *Metals & Alloys,* v. 6, 1935, pp. 307–310.

799. GOUGH, H. J., and D. G. SOPWITH: Some Further Experiments on Atmospheric Action in Fatigue, *J. Inst. Metals,* v. 56, 1935, pp. 55–89.

800. HEMPEL, M., and C. H. PLOCK: Schwingungsfestigkeit und Dämpfungsfähigkeit von unlegierten Stählen in Abhängigkeit von der chemischen Zusammensetzung und der Wärmebehandlung (Fatigue Strength and Damping Capacity of Unalloyed Steels as Dependent on Chemical Composition and Heat Treatment), *Mitt. K.-W. Inst. Eisenforschung,* v. 17, 1935, pp. 19–31.

801. HERTY, C. H., JR.: The Effect of Deoxidation on Some Properties of Steel, *Trans. Am. Soc. Metals,* v. 23, 1935, pp. 113–125.

802. HODGMAN, C. D. (editor): "Handbook of Chemistry and Physics," Chemical Rubber Publishing Company, Cleveland, 20th ed., 1935, p. 919.

803. HUDSON, J. C.: Tests on the Aging of Mild Steel Wire, Iron Steel Inst., *Spec. Rep.* 8, 1935, pp. 191–199.

804. ISENBURGER, H. R.: X-ray Exposure Charts for Steel, *Trans. Am. Soc. Metals*, v. 23, 1935, pp. 614–620.

805. KINZEL, A. B.: Metallurgical Effects Produced in Steel by Fusion Welding, *Trans. Am. Inst. Min. Met. Eng.*, v. 116, 1935, pp. 197–215.

806. LANGEBECK, H. H.: Carbon Loss and Pick-up in the Cupola, *Metals & Alloys*, v. 6, 1935, pp. 31–34, 38 (correlated abstract).

807. MANN, H. C.: Relations of Impact and Tension Tests of Steel, *Metal Progress*, v. 27, No. 3, 1935, pp. 36–41.

808. MCMULLAN, O. W.: Physical Properties of Case-hardened Steels, *Trans. Am. Soc. Metals*, v. 23, 1935, pp. 319–381.

809. MEARS, R. B.: Effect of Composition on the Corrosion Probability of Iron and Steel, *Iron Steel Inst., Carnegie Schol. Mem.*, v. 24, 1935, pp. 69–85.

810. NEWELL, H. D.: The Ductility of Chromium-nickel Austenitic Steels at Elevated Temperatures, *Trans. Am. Soc. Metals*, v. 23, 1935, pp. 225–248.

811. PHILLIPS, A., and H. H. DUNKLE: Directional Properties in Rolled and Annealed Low-carbon Steel, *Trans. Am. Soc. Metals*, v. 23, 1935, pp. 398–408.

812. PIWOWARSKY, E.: Über die Auflösungsgeschwindigkeit des Graphits im flüssigen Eisen (Rate of Solution of Graphite in Molten Iron), *Giesserei*, v. 22, 1935, pp. 274–277.

813. POHL, F.: Beiträge zur Kenntnis des Schalenhartgusses (Chilled Cast Iron), *Giesserei*, v. 22, 1935, pp. 27–31.

814. SAUVEUR, A., and H. L. ANTHONY: Malleable Castings, *Trans. Am. Soc. Metals*, v. 23, 1935, pp. 409–430.

815. SCHULZ, E. H., and W. PÜNGEL: Unlegierte Kohlenstoffstähle (Unalloyed Carbon Steels), "Werkstoff-Handbuch Stahl und Eisen," Verlag Stahleisen m.b.H., Düsseldorf, June, 1935, pp. G-1 to G-6.

816. SHELTON, S. M., and W. H. SWANGER: Fatigue Properties of Steel Wire, *Bur. Stand. J. Research*, v. 14, 1935, pp. 17–32.

817. Society of Automotive Engineers: Iron and Steel Specifications, "S.A.E. Handbook," The Society, New York, 1935, pp. 251–350.

818. TOÚCEDA, E.: Fluidity Test Equipment for White Iron, *Metals & Alloys*, v. 6, 1935, p. 130.

819. VERSÉ, G.: The Elastic Properties of Steel at High Temperatures, *Trans. Am. Soc. Mech. Eng.*, v. 57, 1935, pp. 1–5.

820. WATTS, O. P.: Voltaic Couples and Corrosion, *Trans. Electrochem. Soc.*, v. 67, 1935, pp. 235–257.

1936

821. American Society for Metals: "Metals Handbook," The Society, Cleveland, 1936, 1392 pp.

822. MCKAY, R. J., and R. WORTHINGTON: "Corrosion Resistance of Metals and Alloys," Reinhold Publishing Corporation, New York, 1936, 492 pp.

NAME INDEX

(Item numbers in the Bibliography are in **boldface**; numbers of the pages where the reference is quoted follow the boldface numerals.)

A

B

C

E

F

G

N

T

SUBJECT INDEX

A

B

C

D

E

F

I

L

M

R

S

W